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THE
IRON AND STEEL INSTITUTE
CARNEGIE SCHOLARSHIP
MEMOIRS

VOL. V.

EDITED BY
GEORGE C. LLOYD
SECRETARY



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18...24.

LONDON
E. & F. N. SPON, LIMITED, 57, HAYMARKET
NEW YORK: SPON & CHAMBERLAIN, 123, LIBERTY STREET
SOLD ALSO AT THE OFFICES OF THE INSTITUTE
28, VICTORIA STREET, LONDON, S.W.
1913

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Printed by BALLANTYNE, HANSON & Co.
At the Ballantyne Press, Edinburgh

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THE PRESERVATION OF IRON.¹

BY J. NEWTON FRIEND, D.SC., PH.D. (WORCESTER).

IN his Carnegie memoir for 1912 the present author described a series of "field" tests with large iron plates instituted with the object of determining the relative values of various liquid vehicles and paints capable of being used for the protection of iron and steel. These tests were of two kinds, consisting in exposure to air and to water respectively. In the air tests, plates of wrought iron, No. 22, B.W.G., measuring 18 by 12 inches in area, were cleaned and painted with various mixtures. They were then suspended to a specially constructed wooden fence and exposed to the continued action of the atmosphere. In the water tests, plates of similar iron, measuring 10 by 10 inches in area, were cleaned and painted in the usual way. They were then suspended, each by two glass hooks, in large wooden tanks filled with water and exposed to the open air. At regular intervals of time the surfaces of the plates, in both series of tests, were carefully examined and the observations noted. As the experiments had, at the time when the previous memoir was written, only been in operation for four or five months, the author had to content himself by giving the results of these superficial observations, and drawing what conclusions he could from them.² After the plates had been exposed for from nine to ten months, several of the paint films showed such decided signs of disintegration that it was thought desirable to discontinue the experiments. The plates were accordingly removed, soaked in caustic soda solution to remove the paint, scraped, rubbed clean with old emery paper, and weighed. The loss in weight was taken as measuring the extent of corrosion. The results obtained are given in the following tables.

¹ Received January 10, 1913.

² In an Appendix the results obtained by cleaning and weighing six of the plates were given, and very briefly discussed.

TABLE I.—*Results of Air Tests with Iron Plates painted with various Compositions after Nine and Ten Months' Exposure.*

[Only those results in the separate sections of this table are strictly comparable with one another.]

Plate No.	Date of Exposure.	Treatment of the Plate before Exposure.	Date of Examination.	Observation.	Condition No. ¹	Initial Weight, Grms.	Loss in Weight, Grms.
1	1911 Aug. 28	Clean iron plate without paint.	1912 April 24	Thoroughly corroded.	...	812.7	66.1
2	Do.	Plate painted on each side with 3.3 grams of pure linseed boiled oil.	Do.	The oil film is broken in several places, and slight corrosion of the iron evident over the whole of the surface.	1	815.6	3.5
3	Do.	Plate painted on each side with 3.3 grams of boiled oil containing 0.1 per cent. paraffin wax by weight.	Do.	One or two isolated places where rust has broken through the film. Far less corrosion manifest than in Plate 2.	3	777.7	2.4
4	Do.	Plate painted on each side with 3.3 grams boiled oil containing 0.3 per cent. wax.	Do.	This plate is in a slightly better condition than Plate 3.	4	811.0	1.6
5	Do.	Plate painted on each side with 3.3 grams boiled oil containing 0.5 per cent. wax.	Do.	Best surface of all.	4	799.9	1.6
6	Do.	Plate painted on each side with 3.3 grams boiled oil containing 1 per cent. wax.	Do.	Similar to Plate 4.	3	778.3	1.3
7	Sept. 14	Plate painted on each side with 5 grams of paint consisting of 100 grams dry ferric oxide in 160 grams of pure boiled oil.	July 10	Evidences of slight corrosion under the paint film.	1	812.0	1.0
8	Do.	Exactly similar to Plate 7, but paint contained in addition 0.1 gram paraffin wax per 100 grams of oil.	Do.	Evidence of corrosion less apparent than in Plate 7.	2	824.3	0.5

¹ This number has been arrived at by calling the most disintegrated paint film 1, and grading the others accordingly by their respective *appearances*. No weighings are involved in these observations, which were made before the films were taken down and weighed. The higher the number the better the protection.

TABLE II.—*Results of Small Water-tank Tests with Iron Plates painted with various Compositions after Nine Months' Exposure.*

[Only those results in the separate sections of this table are comparable with one another.]

Plate No.	Date of Exposure.	Treatment of the Plate before Exposure.	Date of Examination.	Observation.	Condition No. ¹	Initial Weight, Grms.	Loss in Weight, Grms.
1	1911 Sept. 15	Plate painted on each side with 2·5 grams of paint containing 100 grams fine rouge in 160 grams of pure boiled linseed oil.	1912 June 16	Same condition as Plate 2.	1	339·17	2·67
2	„ 13	Plate painted on each side with 2·5 grams of paint containing 100 grams rouge (fine and coarse mixed) in 160 grams of pure boiled linseed oil.	June 11	Slight corrosion evident under paint surface.	1	353·2	2·95
3	„ 13	Plate painted on each side with 2·5 grams of mixed rouge paint, same as Plate 2, save that 0·1 gram paraffin wax was added to each 100 grams of oil.	Do.	This plate is in slightly better condition than Plate 2.	2	374·0	3·20
6	„ 14	Rusty iron plate dried at 100° C. and free from scale painted with 2·5 grams mixed rouge paint (see Plate 2) on each side.	Do.	Same condition as Plate 2.	1	313·65	1·11
7	„ 13	Plate painted on each side with 2·5 grams coarse rouge paint containing 100 grams coarse rouge in 160 grams of pure boiled oil.	Do.	Same condition as Plate 2.		358·27	3·62
4	„ 6	Plate painted on each side with 2 grams of pure boiled linseed oil.	Do.	No apparent difference.	...	343·0	2·90
5	„ 6	Plate painted on each side with 2 grams of linseed oil containing 0·1 gram paraffin wax per 100 grams of oil.	Do.		...	355·05	3·75

¹ See note to Table I.

TABLE III.—*Results of Large Water-tank Experiments with Iron Plates after Nine Months' Exposure.*

Plate No.	Date of Exposure.	Treatment of the Plate before Exposure.	Date of Examination.	Observation.	Condition No. 1	Initial Weight, Grms.	Loss in Weight, Grms.
9	1911 Sept. 18	Plate painted on each side with 3·5 grams red lead paint containing 50 grams red lead per 5 grams boiled oil.	1912 June 11	Evidences of slight corrosion under the paint film.	1	308·77	0·27
10	Do.	Plate treated similarly to Plate 9, but the paint contained in addition 0·1 gram paraffin wax per 100 grams boiled oil.	Do.	Same as Plate 10. No difference discernible.	2	335·16	0·16
11	Oct. 4	Plate painted on each side with 2 grams of coarse rouge paint containing 100 grams of coarse rouge per 200 grams of oil.	July 5	Plates 11-16 all in similar condition. Iron slightly corroded under painted surface.	1	379·4	3·85
12	Do.	Plate painted on each side with 2 grams of fine rouge paint containing 100 grams of fine rouge per 200 grams of oil.	Do.		1	372·0	3·20
13	Do.	Exactly similar to Plate 11.	Do.		1	357·5	2·60
14	Do.	Exactly similar to Plate 12.	Do.		1	370·1	3·00
15	Do.	Exactly similar to Plate 11.	Do.		1	363·5	3·00
16	Do.	Exactly similar to Plate 12.	Do.		1	369·65	3·10
17	Oct. 27	Plate painted on each side with 4 grams of mixed rouge paint containing 100 grams of mixed (fine + coarse) rouge per 200 grams of boiled oil.	Do.	Paint surface nearly perfect. Considerably better condition than Plates 11-16.	2	378·2	3·80

¹ See note to Table I.

TABLE III.—*continued*.

Plate No.	Date of Exposure.	Treatment of the Plate before Exposure.	Date of Examination.	Observation.	Condition No. 1	Initial Weight, Grms.	Loss in Weight, Grms.
18	Oct. 27	Rusty plate dried at 100° C., treated in a precisely similar manner to Plate 17.	July 5	Same as Plate 17.	...	367.4	1.85
19	Do.	Exactly similar to Plate 17.	Do.		...	365.28	3.58
20	Do.	Exactly similar to Plate 18.	Do.		...	358.8	2.85
21	Do.	Plate painted on each side with 2 grams of mixed rouge paint (see Plate 17).	Do.	Same as Plates 11-16.	...	360.0	3.10

A careful examination of the above results reveals a number of interesting facts of considerable technical importance.

1. The surface of a paint film does not always give a correct idea of the extent of corrosion taking place beneath. This is particularly clear in the case of Plates 4, 5, and 6 in Table I., and Plates 11, 14, and 17 in Table III. Apparently the only certain method of determining even approximately the extent of corrosion lies in scraping off the paint and examining the surface of the metal beneath.

2. The addition of small quantities of paraffin wax to paint destined for the protection of iron and steel exposed to the air reduces the corrosion very appreciably, as is evident from Table I. If the paint is required to set at the normal rate, not more than 1 lb. of wax should be used for every 1000 lbs. of oil present in the paint. If, however, the rate of setting is less important, the amount of paraffin wax may with advantage be increased to 3 lbs. per 1000 lbs. of oil. This latter is, roughly speaking, equivalent to 2 lbs. of wax per ton of ready mixed paint—an estimate that is sufficiently near for the purpose.

3. In the case of plates immersed in water the addition of the paraffin wax seems rather to assist corrosion than

¹ See note to Table I.

to diminish it; see Table II., Plates 2 and 3, 4 and 5. Plates 9 and 10 (Table III.) apparently lead to an opposite conclusion, but the extent of the corrosion is less here than one-tenth of that in the other plates, and the result is in consequence proportionately less trustworthy.

4. It is interesting to study the effect of using fine and coarse pigments upon the protective power of the resulting paints. Three grades of pigment were used, namely—

- (i) Ferric oxide, as usually sold, consisting of a mixture of fine and coarse particles;
- (ii) Coarse ferric oxide containing particles not less than 0.01 millimetre in diameter, obtained by levigation from (i);
- (iii) Fine ferric oxide, likewise obtained by levigation from (i), consisting of particles less than 0.005 millimetre in diameter.

The middle portion, obtained by levigation of (i), consisted of particles ranging from 0.005 to 0.01 millimetre, and was an exceedingly small fraction of the total pigment used. It was therefore rejected, and pigment (i) was regarded as a mixture of (ii) and (iii).

When we examine Table II. we see that—

Plate 1 (fine rouge) lost	. . .	2.67 grams.
Plate 2 (mixed rouge) lost	. . .	2.95 ..
Plate 7 (coarse rouge) lost	. . .	3.62 ..

Clearly the fine rouge has the advantage, and the mixed rouge yields a mean result between the two extremes.

Table III. yields the following results—

Mean loss of Plates 12, 14, and 16 (fine rouge)	3.10 grams.
Loss of Plate 21 (mixed rouge)	. . .	3.10 ..
Mean loss of Plates 11, 13, and 15 (coarse rouge)	3.15 ..

Here the difference in the protective powers of the three plates is almost negligibly small, but still in the same direction. From this we may conclude that a fine pigment is superior to a coarse one in the preparation of protective paints, a fact which is now generally recognised. It must

be confessed that the results of Table III. were a considerable surprise to the author. The explanation possibly lies in the assumption that pigments prepared by first-class firms at the present time are sufficiently well ground for all practical purposes, and that the result of still further reducing the size of their particles does not very materially increase the value of the pigment. If such is the case, since the cost of grinding greatly increases with the reduction in size of the particles, it will probably prove cheaper to paint a steel structure twice with a good average paint than to protect it once with an excessively finely ground paint. It must be admitted, however, that considerably more work along these lines is desirable before it becomes safe to dogmatise.

5. It is generally believed that two coats of paint preserve an iron surface from corrosion considerably more efficiently than one coat. Liebreich and Spitzer¹ have recently called attention to the fact that in their experiments by increasing the number of coats of paint, not only was the corrosion not reduced, but it actually increased, as if the extra coats stimulated corrosion. Quite possibly this was due to the moisture set free during the oxidation of the paint.²

It will be observed that the results detailed in Table III. lend support to the work of Liebreich and Spitzer. Plates 17 and 19 were painted with two coats of paint on each side at intervals of twenty-four hours. Each coat of paint weighed 2 grams, so that each plate carried twice as much paint as Plate 21.

Mean loss of Plates 17 and 19	.	.	.	3.69 grams.
Loss of Plate 21	.	.	.	3.10 „

6. It is always found that a rusty surface when painted yields unsatisfactory results, and the careful painter removes the rust prior to the application of paint. Undoubtedly this is necessary in damp weather, and it is also necessary to scrape off any scale and thick coats of rust.

It occurred to the author, however, that a slight coating of rust, absolutely free from caking and lumps, might be an

¹ *Zeitschrift für Elektrochemie*, 1912, vol. xviii. pp. 94-99; *Revue de Métallurgie*, 1912, vol. ix. pp. 1-31, 78-111, 160-186, 275-293.

² See the criticism of K. Arndt, *Zeitschrift für Elektrochemie*, 1912, vol. xviii. p. 233.

advantage rather than otherwise, if perfectly dry, for two reasons :—

1. Fine rust is practically a pure ochre, very free, in the case of wrought iron and steel, from silica. Hence the surface is very absorbent, and if painted with a good coating of thin paint a similar result may be expected as by first laboriously cleaning the metal and then painting with a thicker paint—that is, one containing more pigment in proportion to oil than the former.

2. If the iron surface is sandpapered down it is made much smoother, and its area is reduced to a minimum. If, however, the rust is left on as a thin coating, the underlying surface of metal in contact with the rust is slightly pitted and uneven. Hence when the paint is applied it has a larger surface to grip, and its tenacity will therefore be increased in proportion.

Five plates were allowed to rust by exposure to weather, and when a thin layer of rust had been formed (after about a week), the plates were thoroughly dried in a steam oven at 100° C., weighed, and three of them finally painted with rouge whilst the rust was still adhering. (See Plates 6, 18, and 20.)

The remaining two plates served as blanks, being weighed, then cleaned, and again weighed, the loss in weight giving the amount of rust that had accumulated during the week's exposure. When Plates 6, 18, and 20 had been exposed, after painting, for a sufficient length of time in the tanks, they were carefully cleaned and weighed, the loss in weight giving the total amount of corrosion. Subtracting the weight of rust already present on the plates when they were painted, as calculated from the two blanks, the difference gave the corrosion which had taken place after the application of the paint. On comparing Plate 2 with 6, 17 with 18, and 19 with 20, it will be observed that the plates suffered far less from corrosion when painted in a rusty condition than when first polished bright before painting. This is a remarkable result.

It would seem, therefore, that when the painter is called upon to protect by painting thin sheets of iron, such as galvanised iron, which has already rusted somewhat, any

lumps of scale should be removed, and a coat of paint applied in warm dry weather to the rusty surface. The danger of the sheets rusting through will thus be considerably less than if the rust is first carefully scraped off and the metallic surface polished with sandpaper before the application of paint, as is the more usual custom. In the case of structural work, iron palings, &c., where the painting is as much for decoration as protection, and where the iron is so thick that the loss of a little as rust is immaterial, it is probably better to polish the metallic surface before painting, as the appearance is thereby greatly improved.

From the above remarks it will be clear that a great deal of work yet remains to be done in this important field. The results contained in this memoir are not to be regarded as final, but rather as indications of the lines along which useful research might be carried out.

The author wishes to acknowledge the kindness of the Carnegie Scholarship Committee who awarded him a grant in May 1911, which has enabled him to obtain the results detailed above.

THE DEVELOPMENT OF THE DRILL TEST AS A MEANS OF ASCERTAINING THE MACHINING PROPERTIES OF IRON AND OTHER METALS AND FOR THE INVESTIGATION OF TOOL STEELS.¹

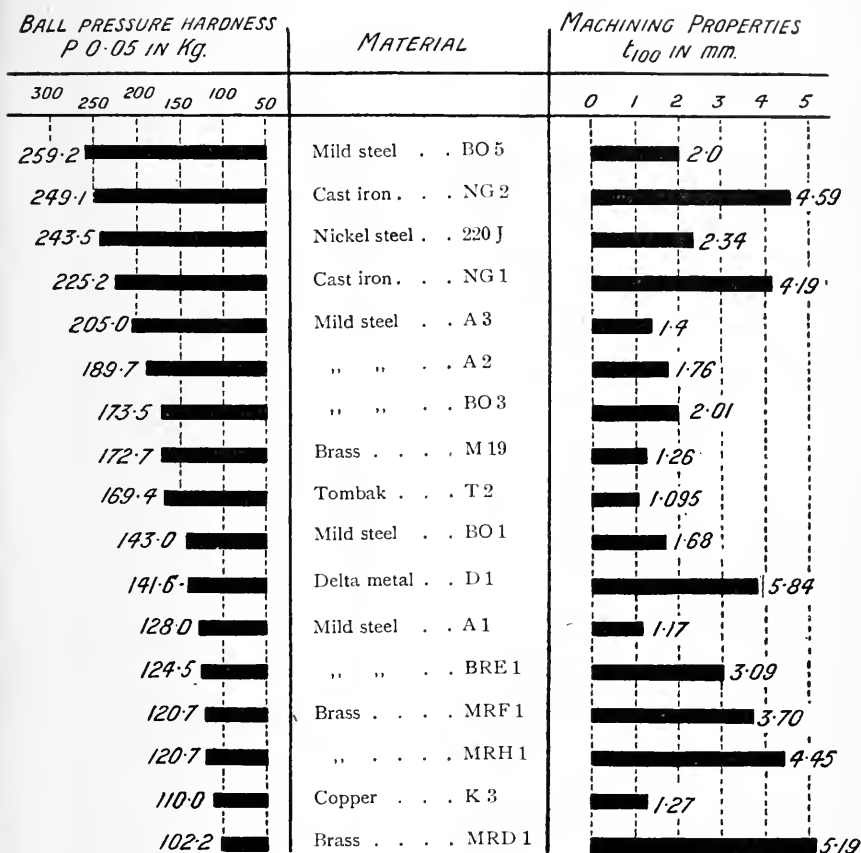
BY A. KESSNER (KÖNIGLICH TECHNISCHE HOCHSCHULE, BERLIN).

I. INTRODUCTORY.

THE present investigation is intended to show in what way an objective test of the machining properties of metals may be undertaken with a simple testing apparatus. The practical value of such a test is now fully recognised, and the experiments made in this direction in recent years by several investigators have been followed with interest, though they have not as yet led to the evolution of a suitable practical method. Since the introduction of turret lathes, and especially automatic turret lathes, the utility of which depend chiefly upon their capacity for taking the heaviest possible cut at one time, the objective testing of the machining properties of metals has become a serious question, the solution of which is the object of the present study. When it is further considered that by far the greater number of castings and forgings used in machine construction receive their final form under a cutting machine tool of some kind, and that the cost of machining is many times more than that of the production of the piece in the rough, it will be conceded that the machining properties of a piece cannot be passed over as negligible. In general, it may be said that the working time and wages spent on machining, apart from the time required for setting and taking down, depend essentially upon the machining properties of the material.

The machining properties are therefore a governing factor in the cost of the finished article.

¹ Received February 27, 1913.



Ball pressure hardness P 0.05 is the pressure in kilogrammes required to produce an impression 0.05 millimetres deep with a steel ball 5 millimetres in diameter.

Machining properties t_{100} = the depth of hole attained with 100 revolutions of the drill at equal pressure on the drill.

Flat drills 8 millimetres diameter.
Angle ϕ of the point = 110° .
Clearance angle $\alpha = 15^\circ$.
Pressure on drill P = 56 kilog.

FIG. 1.—Comparison between Ball-pressure Hardness and Machining Properties of various Materials.

All efforts to deduce from the ball-pressure hardness the measure of the machining properties must fail, since the two characteristics are not in any way comparable with each other. Thus experience shows that aluminium is more difficult to machine than mild steel, although the ball-pressure hardness of aluminium, according to Brinell, is only 38, whilst that of mild steel is about 100. Another example of the incomparability of ball-pressure hardness with the machining capacity is afforded by cast iron test-pieces of different sections cast from the same ladle; also by brass test-pieces ($\text{Cu} : \text{Zn} = 2 : 1$) with an increasing proportion of lead. Tests of this kind are described on pages 37-44 and Fig. 1.

Reasoning on these lines, Heyn has deduced the following law:¹—

“Machining capacity under cutting tools depends not only on the hardness (ball-pressure hardness, for instance), but also on the toughness. Great hardness as well as great toughness render the machining more difficult.”

All experiments with which the author is acquainted for ascertaining the machining properties of metals are based on the drill test. The deeper a drill, running at a given number of revolutions and under constant external load, penetrates into the material within a definite time, the more readily can the material be machined. The depth reached after a certain number of revolutions is therefore a measure of the machining properties. All appliances for testing such properties must therefore be provided with means for registering the revolutions, and the depth to which the hole has been drilled. This is best done by the automatic recording of diagrams. Thus the testing machines to be described later automatically trace a diagram on which is plotted the depth of the hole on an enlarged scale. The depths thus enlarged are plotted as abscissæ, and the number of the revolutions of the drill as ordinates.

The idea of using the drill as a means for testing the machining properties was first conceived by C. A. Bauer,² who

¹ Martens-Heyn, *Materialienkunde für den Maschinenbau*, vol. II A. p. 414.

² C. A. Bauer: “A Novel Method for Testing Cast Iron for Hardness” (*American Machinist*, 1897, p. 245).

described his method and machine in 1897. The drilling test was further investigated by Keep,¹ Leyde,² and Reininger, but none of these observers succeeded in finding a satisfactory solution of the problem.

It may be noted here that such drill tests never can yield absolute values, but only comparative values obtainable under similar conditions. These comparative values are got by first drilling a standard material, and afterwards, under exactly the same conditions, the material to be tested, and comparing the depths of the holes in each case.

The first difficulty is to find a material suitable as a standard, as it must be one which can be reproduced at any time with the same chemical composition and physical properties. The above-mentioned investigators used cast iron as a standard material for comparison, but the author's experiments led him to the conclusion that of all materials cast iron is the most unsuitable one that can be imagined.

The design, construction, and standardising of suitable testing apparatus present further difficulties, to the surmounting of which the author specially devoted his attention. The programme of his work was laid down as follows:—

1. Construction of an apparatus which can be fitted to any drilling machine, and will automatically trace a diagram upon which are recorded the revolutions of the drill and the depth of hole, the external load on the drill being maintained constant.
2. The calibration of the apparatus designed to accomplish these purposes.
3. Experiments for determining the most suitable form of drill and best angle of cutting edge.
4. Experiments for determining the proper load to apply to the drill and the best peripheral speed of the drill.
5. Experiments on the choice of a standard material.
6. Practical examples of the application of the apparatus described.

¹ Keep: *Iron Age*, 1899, p. 9; *ibid.*, 1900, p. 16. See also *Stahl und Eisen*, 1910, p. 1010.

² O. Leyde: "Die Prüfung des Gusseisens" (*Zeitschrift des Vereins Deutscher Ingenieure*, 1904, p. 169; *Stahl und Eisen*, 1904, p. 186).

II. THE DIAGRAM OF THE DRILL TEST.

The depth of hole reached after a certain number of revolutions is recorded on the diagram of the drill test. Revolutions of the drill are automatically traced as ordinates by the pencil of the indicator and the depths of hole as abscissæ. Turning to diagram (Fig. 2) it will be seen that in the case of Material I. the drill has penetrated to a depth of 5 millimetres, while with Material II. it has gone to a depth of 10 millimetres, the revolutions in both cases being 100. If the material is homogeneous and the external

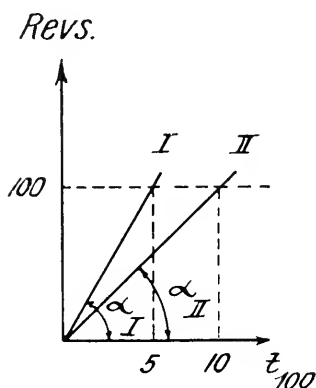


FIG. 2.

conditions remain unchanged, the drilling curve follows a straight line. The angle α , at which this curve inclines to the horizontal, is the measure of the machining properties of the material. According to Fig. 2, therefore, Material I. is more difficult to machine than Material II.

The slightest deviation of the curve from the straight line indicates either irregularities within the material itself, or a variation in the external conditions. The latter are affected by the shape of the drill, the state of the drill steel, the cutting angle of the drill, the means for removing the drillings, the friction against the sides of the hole, the load on the drill, the cutting speed, and, finally, the elastic deformations in various parts of the indicator.

Assuming that these external conditions remain unchanged during the test, the following conclusions may be drawn from Fig. 3.

Material III. has at h a hard spot, and Material IV. at w a soft spot (a blowhole or something of that nature). Material V. is harder towards the inside, and Material VI. softer.

From the course of the curves it is possible to detect every kind of change in the external conditions. For instance, in Material III. the deflection of the curve at h can be attributed to the jamming of the cuttings under the drill. The course

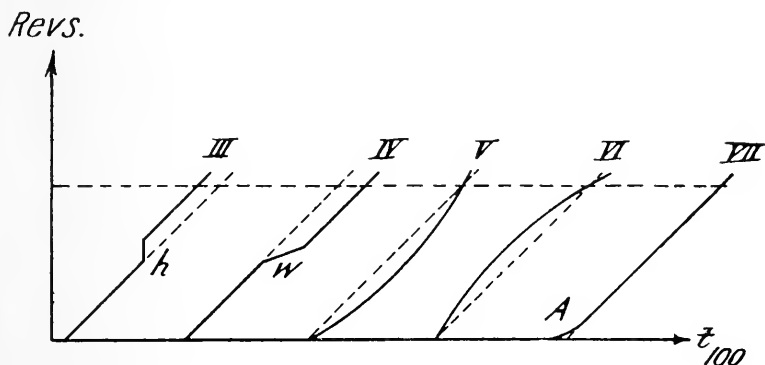


FIG. 3.

of the curve of Material V. indicates an increasing amount of friction in the hole and an increasing elastic stretching of the indicator cord. An often noticeable feature is the form taken in the curve marked VII., which is due to the irregular attack of the drill at starting. The "start" of the drill (Fig. 4)—that is, the vertical distance through which the point of the drill must descend until both cutting edges attack equally—is denoted by A. Similar irregularities, as manifested in the actual operation of tests, will be discussed later.

In view of all these circumstances which may cause the curve to deviate from the straight line, the author decided to replace the drilling curves III., IV., and VII. by a dotted straight line. In those cases where a well-defined bending

took place, as in curves V. and VI., a straight line is not plotted, but the actual depth of the hole has been measured.

The depth of hole attained after 100 revolutions of the drill is to be taken as the standard measure of the machining

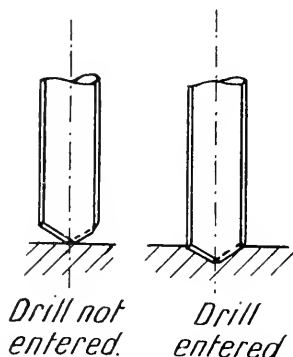


FIG. 4.

properties, and for the purpose of convenience in all the following experiments this depth is denoted simply by the term t_{100} . It is of course to be understood that all the comparative tests were performed under constant external conditions—that is, the load on the drill, the speed of revolution,

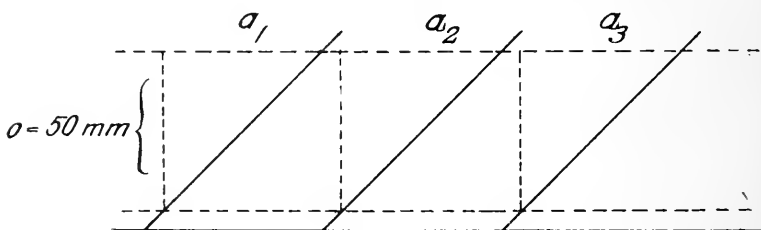


FIG. 5.

the drill diameter, and the dimensions and angle of the cutting edges remained the same.

In all three models of the testing machines about to be described the vertical travel of the pencil is so controlled that 1 millimetre length of ordinate corresponds to 2 revolutions of the drill spindle (Fig. 5). All that is therefore required after

removal of the diagram from the indicator is to draw two lines parallel to the base line and to measure on the abscissæ a_1, a_2, a_3 , &c., the depths of the holes on a definite enlarged

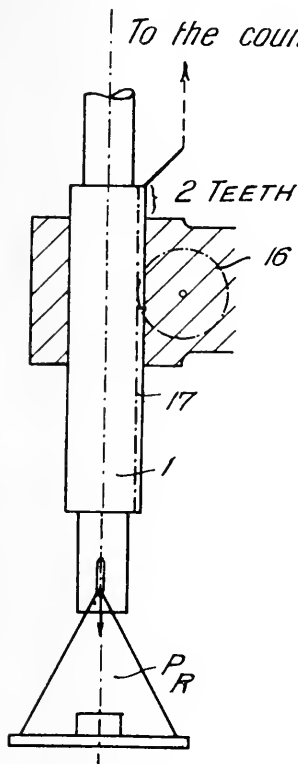


FIG. 6.

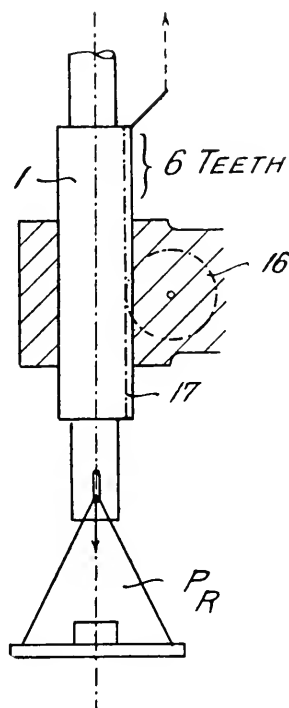


FIG. 7.

scale. The enlargement with each of the three machines used was accurately determined by means of measuring instruments and amounts to—

For Model	I.	4.92 millimetres.	Length of abscissa=	1 millimetre depth of hole.
" "	II.	4.75 "	" "	=1 " " "
" "	III.	8.7 "	" "	=1 " " "

III. THE TOOLS FOR TESTING THE MACHINING PROPERTIES.

A. *The Hardness Drilling Tool. Pattern I. (Kessner Type.)*

At the instance of Professor Heyn, the author, in 1908, designed an apparatus for testing the machining properties of metals which can be fitted to any vertical drilling machine without withdrawing it from service in the workshop. Equipped with this apparatus, small workshops which have no laboratory fitted up with expensive special tools can exercise continuous control over the cutting capacity of their tools and the machining properties of the metals to be operated upon. In practice it may often be convenient to compare the machining properties or homogeneity at different parts of a large casting, and by means of this apparatus the drill tests can be made in the workshop, and as nearly as possible at the very part where holes are required to be drilled. In such cases it must be an obvious advantage for the drill test and the ordinary process of drilling holes to be performed upon one and the same machine.

The principal component parts of this hardness drilling tool are illustrated in Fig. 8, which shows the screw spindle 4, with the nut 5 carrying the indicator pencil 7, the indicator drum 10, and the loading lever 13. The revolutions of the drill spindle 1 are transmitted by gear wheels 2 and 3 to the screw spindle 4, on which the nut carrying the pencil moves between lateral guides 6. The pencil 7 is lightly pressed against the indicator drum by means of the spring 8 and adjusting screw 9. The gear ratio is such that the pencil travels through a vertical distance of 1 millimetre for every 2 revolutions of the drill spindle. The screw spindle, the guide bars, and the indicator drum are attached to a plate 11, which is clamped to the frame of the drilling machine.

The loading lever 13, consisting of two segments 14 and 15, is slipped on to the free end of the pin 12, which forms part of the usual hand-feed gear. The pinion 16, which is keyed on to the pin 12, catches in the teeth of the rack 17 fixed on the drill spindle 1, so that the weight 18 attached to

the segment 14 applies a constant external load to the drill

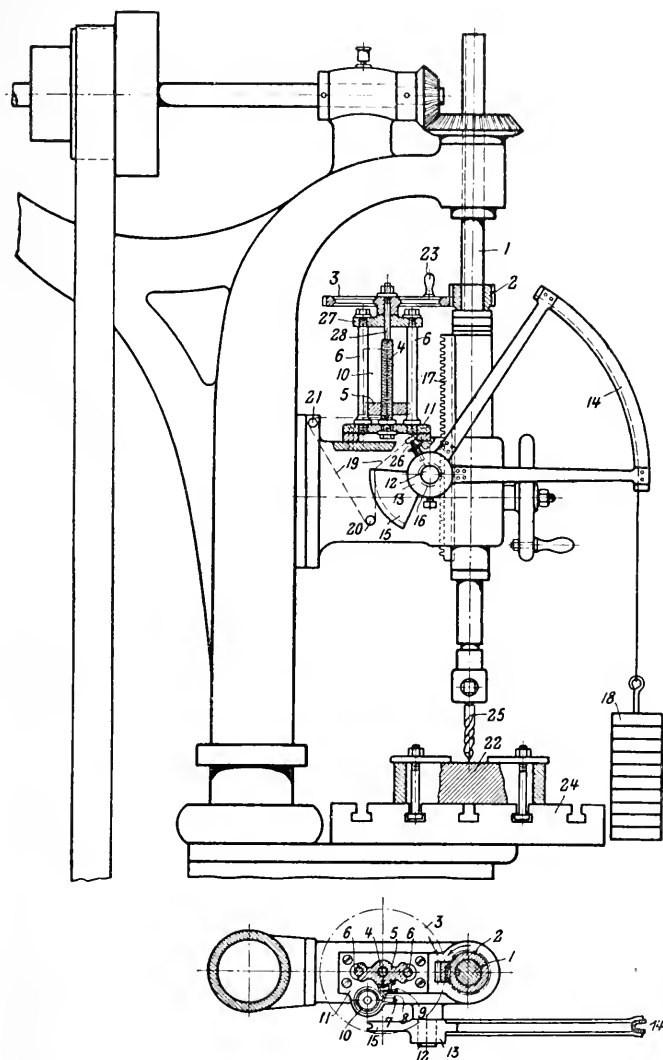


FIG. 8.—Hardness Drilling Machine. Pattern I. (Kessner Type.)

spindle. The motion of the spindle in descending is transmitted by a thin cord or steel wire 19, passing over segment

15 and the guide pulleys 20 and 21 to the indicator drum carrying a sheet of paper, which is thereby caused to revolve. After the test the drum is brought back to its original position by means of a spring. The fine adjustment screw 26 enables several curves to be traced at convenient distances apart on one sheet. The ratio of the movements is such that 1 millimetre depth of hole in the test-piece 22 corresponds to an abscissa length of 5 millimetres in the diagram—that is, the depth is shown magnified 5 times.

At the commencement of the test the drill spindle 1 is lowered until the spur wheels 2, 3, are thrown out of gear. By turning the handle 23 the nut 5 is brought to its lowest position and the pencil is adjusted, so that the point just touches the drum lightly. The drill spindle is now raised until the spur wheels are again put into gear, the ends of the teeth being filed to a point in order to facilitate their sliding in. The table of the machine with the test-piece mounted on it is then adjusted so that the drill touches the piece on which an entry has previously been made with another drill with a point of the same angle. The load lever 13 should be adjusted to the position shown in Fig. 8 in order to utilise the arc of the segment as fully as possible. Finally, the weight 18 is loaded up and the indicator drum is set by means of the fine adjustment screw into the proper position for taking up the diagram. When the machine has been started the pencil will trace the drilling curve resulting from the rotation of the spindle and the vertical distance through which it descends.

In order to prevent the nut 5 from running up and jamming against the upper cross-piece supporting the screw spindle 4 at the end of the test, the screw thread is turned off at the upper end of the spindle for a distance sufficient to allow the nut to run clear of the thread. When the tests are finished the drilling machine can again be put in order for ordinary work by the removal of the spur wheel 3, and of the load lever 13, and the refixing of the handle on pin 12 for the hand feed.

THE HARDNESS TESTING DRILLING MACHINE.

Pattern II. (Kessner type).

The hardness drilling machine, according to Pattern I., having been completed and tested, the necessity was manifested for a type of apparatus for testing the machining properties of metals which could be attached to a lighter drilling machine. In designing Pattern II. of the hardness drilling machine, some improvements were made, of which the following are principal ones:—

1. The indicator is made self-contained in one case, which is fixed on the side of the two upper guide brackets, 36, 37, about level with the eye (see Fig. 32, Plate IV.).
2. The driving spur wheel can be moved out of gear by means of lever 2*a*.
3. The screw spindle 4 at its lower end runs on a ball-bearing.
4. Of the two guide columns only one (6) is retained.
5. The nut 5 is slotted, and is made adjustable by means of screw 5*a* (Fig. 31).
6. In place of the segment 14 of the load lever the feed-wheel 14*a* is fitted, which is fixed solid on the end of the pin carrying the pinion (Fig. 9, Plate I.).
7. A pulley, 15*a*, takes the place of segment 15. This pulley, in the groove of which the cord 19 runs for rotating the indicator drum, revolves loose on the hub of the feed-wheel, but is clamped tight to the wheel by the screw 15*b* when the machine is in operation. This pulley does away with the need for the fine adjustment 26 in Pattern I., because it can be clamped to the feed-wheel in any position, and so enables any number of curves to be traced one above the other on the indicator diagram.

Nearly all the tests performed up to the end of December 1912 were carried out with these hardness drilling machines, Patterns I. and II., at the Mechanical Technical Laboratory of the Royal Technical High School, Berlin. It became apparent at the outset that, notwithstanding the fact that the curves

usually followed a straight line course, the values obtained showed many unaccountable discrepancies which involved frequent repetition of the experiments. The author attributed these numerous failures to the frictional resistances within the machine itself, and especially to the want of accuracy in the contact between the pinion 16 and the toothed rack 17. The correctness of his supposition was proved by subsequent investigations, the substance of which was as follows:—

In the case of the hardness testing drilling machine, Pattern II., the force P_R required to pull down the drill spindle 1 was determined by suspending weights from it while the machine was running at 710 revolutions per minute, the spindle being counterbalanced (as shown in Figs. 6 and 7). The force P_R was found to vary considerably according to the position of the spindle. The results are entered in the following table (each value being the average of ten tests). P_R represents the weight in kilogrammes, and 2Z, 3Z (meaning that 2 teeth, 3 teeth, &c., of the rack projected above the bearing), represent the position of the drill spindle.

Position of drill spindle	.	.	2Z	3Z	4Z	5Z	6Z
P_R in kilogrammes	.	.	7.2	11.8	9.6	9.6	9.3

The author therefore considers that the discrepancies were attributable: (1) to the unequal friction in the guide bearing; (2) to want of accuracy in the engagement of the pinion (with 18 teeth) in the rack 17. These considerations led to the reconstruction of the machine according to Pattern III.

Having determined the amount of the force P_R , the drill spindle of the machine was always adjusted to the same height at the commencement of all subsequent tests, the position being that shown in Fig. 6.

Pressure on the Drill.

For determining the effective pressure on the drill the arrangement shown in Figs. 10 and 11 was planned. In place of the drill a plain round tool with a hemispherical end was fixed. This tool was brought down upon a well-lubricated cast iron plate, 22a, recessed to take the round end, and the

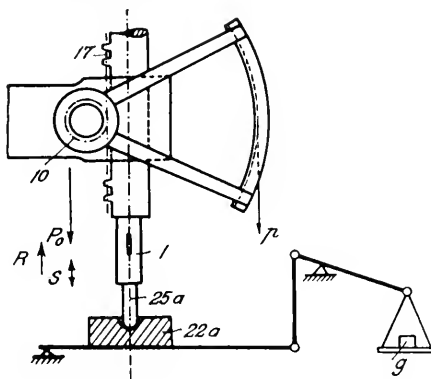


FIG. 10.—Determination of the actual Drill Pressure P with the Hardness Drilling Machines. Patterns I. and II.

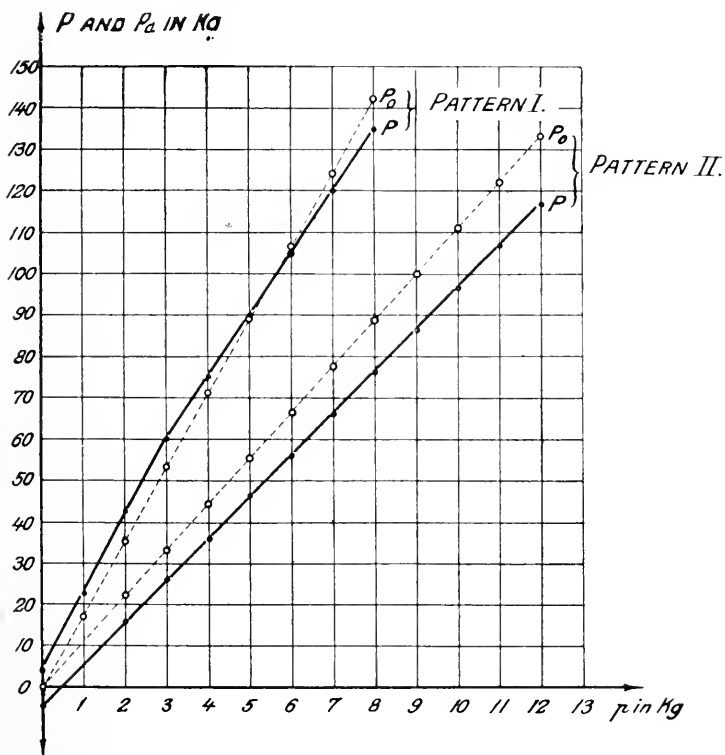


FIG. 11.

pressure upon the plate was transmitted to a lever balance. The machine of Pattern I. was driven at 228 revolutions, and that of Pattern II. at 710 revolutions per minute.

At the beginning of the test the scale pan was loaded until it assumed the position shown in Fig. 10. The weights were then gradually removed from the pan until the drill spindle began to move downwards. The forces acting on the spindle were then:—

1. Downwards. The theoretical pressure P_0 due to the load on the segment 14 (Pattern I.) or on the feed-wheel, 14 α (Pattern II.).

2. Upwards. The total friction R opposed to the direction of the movement of the spindle.

3. Upwards or downwards. The unbalanced portion S of the weight of the spindle.

In this test the weight g , weighing down the scale-pan of the decimal lever balance, corresponds to the effective pressure P on the drill.

That is, $P = 10g = P_0 \pm S - R$.

The theoretical pressure on the drill P_0 for the two machines was found to be

For Pattern I., $P_0 = 17.75 P$.

„ „ II., $P_0 = 11.11 P$.

With the aid of the decimal lever balance, the effective pressure P on the drill was then determined for the different degrees of load p .

TABLE I.

Kilo- grammes, p .	Pattern I.		Pattern II.	
	P_0 .	P .	P_0 .	P .
0	0.0	4.1	0.0	4.6
1	17.5	23.0
2	35.5	43.0	22.2	16.6
3	53.3	60.0	33.3	26.2
4	71.0	75.0	44.4	36.2
5	88.8	90.0	55.6	46.6
6	106.5	106.0	66.7	56.3
7	124.0	120.0	77.8	65.9
8	142.0	135.0	88.9	76.3
9	100.0	86.5
10	111.1	96.7
11	122.2	106.9
12	133.3	116.6

The values given in Table I. are plotted in the curves presented in Fig. 11.

The theoretical pressure P_0 on the drill must pass through the starting-point of the co-ordinates, and should follow an upward straight line course. With the machine of Pattern I., at low loads ($p = 1-5$ kilogrammes) the effective pressure P on the drill is greater than the theoretical pressure P_0 , which proves that the counter weight is not able to counterbalance the weight of the spindle exactly. Only when the load increases until p is greater than 5 kilogrammes, does P gradually fall below P_0 , this being attributable to the increase in the amount of friction in the spindle bearing. Owing to the inaccurate contact of the pinion in the teeth of the rack a component force arises which presses the bush of the drill spindle hard against the guide bearing. By this action the friction in the bearing is increased, and the spindle is no longer sufficiently counterbalanced, owing to being held more firmly in the bearing.

In the case of the machine of Pattern II., the friction conditions are more favourable than in that of Pattern I. The curve of the theoretical pressure P_0 on the drill at the outset lies higher than the curve of the effective pressure P . When $p = 0$, $P = -4.6$ kilogrammes, thus proving that the counter weight exceeds the weight of the spindle. But as the load p increases the difference between P_0 and P becomes greater, which again is due to the increasing friction in the spindle bearing.

Shape and Cutting Angle of the Drill.

The choice of a suitable drill is a matter of great importance in testing the machining capacity of materials. The author's first attempts were made with twist drills, which were re-ground after every hole on a special grinding machine for twist drills. It was soon found, however, that it was absolutely impossible to avoid slight displacements of the drill in the carrier of the grinding machine, with the consequence that the test results showed considerable discrepancies. Also, any slight alterations, of apparently the most trivial kind, in the adjustment of the grinding machine influenced the results beyond all control.

Another drawback of the twist drill was the uncertainty of being able to grind the cutting edges exactly to the required angle. In view of these disadvantages, to which may be added the great expenditure of time in re-grinding, the author sought a drill of a more suitable shape.

The simplest form which presented itself was a flat drill, as shown in Figs. 12 to 14, which could be ground to any desired angle on an ordinary workshop grinder, thus dispensing with the need for a special grinding machine. It thus became possible to change at will the angle of the point, the angle β of the breast, and the clearance angle α , until the most suitable shape was attained. The two first-named angles, ϕ and α , could be adjusted without any difficulty on an ordinary grinding wheel, but in order to obtain the best angle β of the breast of the drill, careful calculation was required. According to Professor Heyn, the effective breast angle β is defined in the following way:—

A plane, which is vertical to the clearing surface and parallel to the direction of the cut, intersects the surface of the tool breast in a line a and the clearing surface in a line b . The angle β formed by the two lines a and b is the effective breast angle.

The clearance surface itself is defined as the surface described by the cutting edge during its movement in cutting operation. The following formulæ have been deduced by Professor Heyn for the calculation of the effective breast angle.

(1) For the two cutting edges AC:—

$$\cotan \beta = -\sin \rho, \tan \epsilon$$

$$\tan \rho = \frac{AC \cotan \frac{\phi}{2}}{\frac{D}{2} - OA}$$

in which D = diameter of drill.

(2) For the connecting ridge AOA:—

$$\tan \beta = \frac{-\cotan \frac{\phi}{2}}{\sin \epsilon}$$

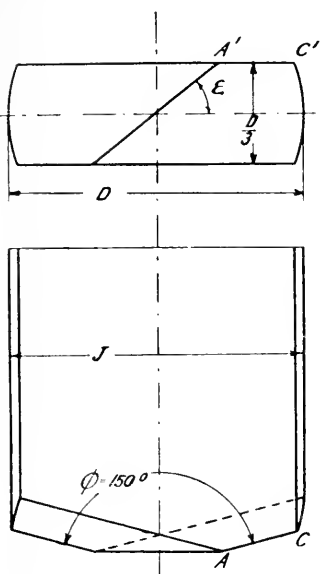


FIG. 12.

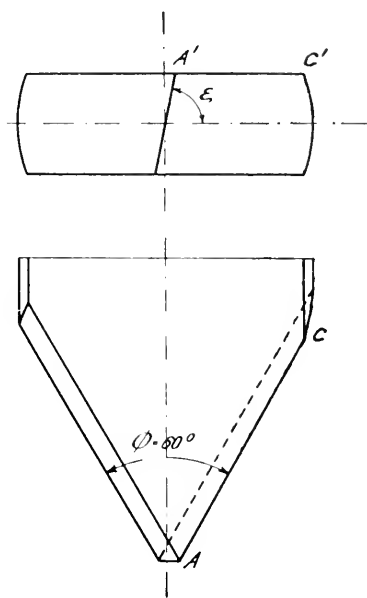


FIG. 13.

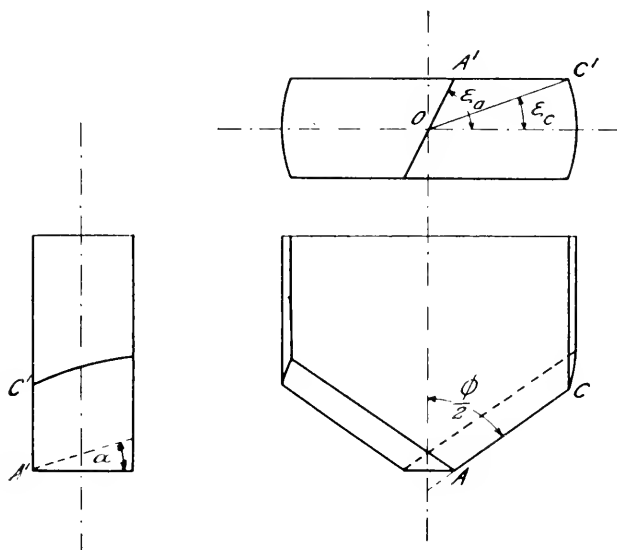


FIG. 14.

According as the angle β of the breast is calculated at A, at C, or at O, different values are obtained, which are shown graphically in a curve in Fig. 19. The effect of the several angles on the depth attained in drilling is discussed further on.

From numerous tests made with the flat drills shown in Figs. 12 to 14, it was found that the connecting ridge AOA soon became blunted, until the author hit upon the plan of shortening the ridge by grooving the flat faces of the drill, as shown in Fig. 16. The form of drill there represented proved to be the most suitable, and these drills stood better than any other in all subsequent tests. The author therefore considers that this is the best type of drill for hardness drilling tests. After once grinding, the drill would perform 25 to 30 tests without any noticeable blunting, and in the experiments to be described later the drill required to be ground only once for every 20 to 30 holes drilled. The breast angles of the drill used for the tests, the results of which are plotted graphically in Fig. 29, gave even better values at the grooved parts than those of the flat drills represented in diagram Fig. 19, as may be clearly judged from the angle ϵ in Fig. 14. By shortening the connecting ridge AOA by one-half, the unfavourable effect of the much too obtuse angle β_v is greatly reduced, and there is no fear of weakening the drill too much by milling the grooves along the centre of its faces, as it is not subjected to any excessive stress.

The small amount of friction of the two cylindrical sides of the drill against the walls of the hole, which seldom exceeds 1 to 6 millimetres in depth, does not appreciably affect the result. Even when a second test was performed in the same hole, that is, by deepening the first hole, nothing on the diagram was observed that could be attributed to side friction of the drill. It was therefore of no advantage to make the sides of the drill taper upwards towards the shank, as this would only have led to the old trouble with the pointed drill, where the diameter is reduced each time the tool is re-ground. The cuttings from the new flat drill also presented no appreciable hindrance, as they work up out of the hole in the ordinary course.

It is particularly to be noted that this flat drill, as shown in

Fig. 16, always gives a greater depth of hole than a twist drill, given the same diameter and the same pressure on the drill. It therefore greatly excels the latter in cutting capacity. Finally, it may be mentioned that the numerous attempts of the author to eliminate the unfavourable cutting effect of the connecting ridge AOA by first boring a small hole of a diameter equal to the length of the ridge failed to give any satisfactory result.

The influence of the clearance angle α on the machining capacity of some materials is shown in Fig. 18, which represents the average values of 250 tests. Flat drills as described by the author were used with angle $\alpha = 5^\circ$ to 30° , and the angle ϕ of the point $= 130^\circ$, this latter angle being chosen because

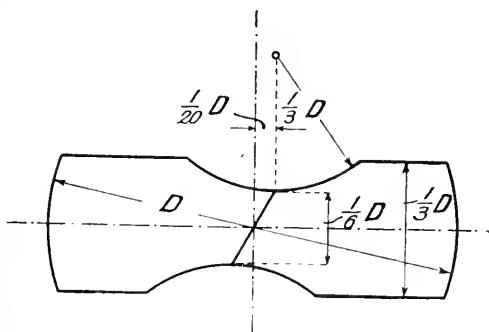


FIG. 16.—Form of Drill finally adopted.

the author had previously obtained satisfactory results therewith.

The explanation of the phenomenon that the machining capacity at first increases as the clearance angle increases is no doubt to be found in the fact that when α is larger the drillings under the cutting edge of the drill can escape more easily. The jamming of the drillings under the drill is indicated in the diagram by the hump at h (see Fig. 3). The drill continues to revolve, but for a time the depth of the hole is not increased.

With the mild steel A2, t_{100} (depth of hole in millimetres obtained after 100 revolutions) equals 0.85 millimetre with angles $\alpha = 35^\circ$, and it rose to 1.63 millimetre with angle $\alpha = 30^\circ$, that is, t_{100} was nearly double. With cast iron NG₄,

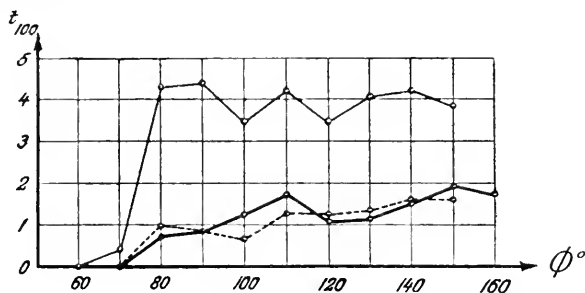
under the same conditions, the corresponding values for t_{100} were 2.76 and 5.81 millimetres. With copper an increase in the value of t_{100} was noticeable only between clearance angles of 5° and 10° , and it varied very little when α was increased beyond 10° . From this it may be concluded that with angle ϕ of the point of the drill $= 130^\circ$, the variation in the clearance angle α has the greatest effect in the case of cast iron and the smallest in that of copper. It was especially noteworthy that in ten tests on delta metal the drill with clearance angle $= 5^\circ$ would not penetrate at all; and in a further series of ten tests on the same material with angle α increased to 10° , $t_{100} = 1.3$ millimetre. With α rising from 10° to 30° the curve for delta metal follows a similar course to that of the cast iron curve.

The most uniform results with all these materials were obtained when $\alpha = 15^\circ$ to 20° , for which reason that angle was regularly adopted in all subsequent experiments. The influence of the angle ϕ of the drill point on the machining capacity of some materials was determined with the author's flat drills, using a constant clearance angle $\alpha = 15^\circ$. Six hundred tests were made altogether, the drill running at 710 revolutions per minute, and the pressure being 56.3 kilogrammes.

The results of these tests are summarised in curves in Fig. 17, which represent the average of 580 tests. The tests show that with cast iron NG₄ the drill with angle $\phi = 60^\circ$ does not penetrate at all. The angle of the breast at the cutting edges varies from 110° to 168° , and at the connecting ridge $= 119^\circ$. With $\phi = 70^\circ$ and consequently $\beta = 109^\circ$ to 166° , $\beta_v = 124^\circ$ at the connecting ridge, t_{100} did not exceed 0.42 millimetre—a very low value. Strange to say, however, the drill pointed to an angle $\phi = 80^\circ$, corresponding to an angle $\beta = 108^\circ$ to 161° , and $\beta_v = 120^\circ$ at the cutting edges, penetrated to a depth of $t_{100} = 4.3$ millimetres. With the angle ϕ increasing to 150° , the value of t_{100} remains practically the same, except at the points $\phi = 100^\circ$ and 120° , at which a drop is recorded, due to the alteration of angle β , which is explained later.

With mild steel A2, and also with copper K3, the drill

pointed to angles $\phi = 60^\circ$ and 70° does not penetrate at all, its cutting effect beginning only when ϕ is increased to 80° . As ϕ increases t_{100} at first increases in the case of the mild steel A2, and reaches the value of 1.75 millimetre with $\phi = 110^\circ$, then drops to 1.1 millimetre with $\phi = 120^\circ$, and rises again to a maximum of 1.95 millimetre with $\phi = 150^\circ$. These variations bring the author back to the consideration of the various sizes of the breast angle β as affected by increasing the angle ϕ . When ϕ is small, the cutting edges are long, the connecting ridge is short, and the average angle β at the cutting edges is less favourable. When ϕ is large the opposite is the case. The reduction of the length of the



Clearance angle $\alpha = 15^\circ$; P = 56.3 kilogrammes; 710 revolutions per minute.
Mean of every 20 tests (580 tests in all).

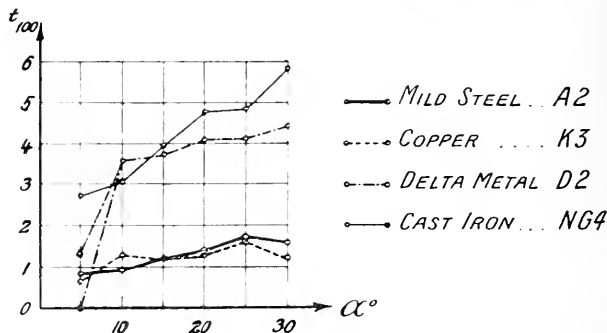
FIG. 17.—Influence of Angle ϕ of Point on the Value t_{100} of various Materials.

connecting ridge in the manner devised by the author, as shown in Figs. 15 (Plate II.) and 16, increases the efficiency of the drill in any case.

The most regular results were obtained when $\phi = 110^\circ$ and 150° . The latter size of angle proved unsuitable, however, as the drill did not centre well on account of the relatively long connecting ridge and short cutting edges (see Fig. 12). The author therefore chose, as the result of these and the previous experiments, a drill pointed to angle $\phi = 110^\circ$ and clearance angle $\alpha = 15^\circ$ to 20° , as being the most suitable form.

The influence of the pressure P on the drill on the machining properties of cast iron and mild steel is shown in Fig. 20. The tests were made with the hardness drilling tool, Pattern II., and with a flat drill 8 millimetres diameter (Fig. 16), in

which angle ϕ of the point = 130° and the clearance angle $\alpha = 15^\circ$. The two curves show that an increased drilling



Angle ϕ of point = 130° ; $P = 56.3$ kilogrammes; 710 revolutions per minute.
Mean of every 10 tests (250 tests in all).

FIG. 18.—Influence of Clearance Angle α on Value of t_{100} of various materials.

pressure has a much greater effect on cast iron than on mild steel. At low loads with $P = 25$ to 46 kilogrammes, the indi-

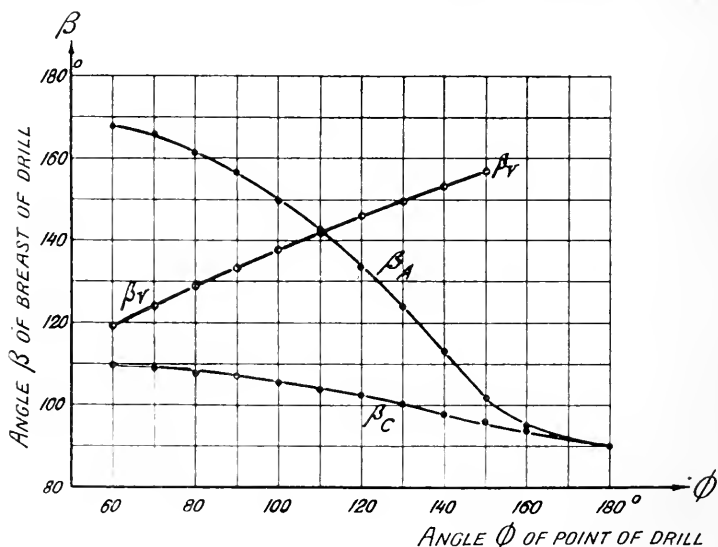


FIG. 19.

vidual drilling curves frequently show a break (similar to Material III., Fig. 3), which is caused by the jamming of the

drillings under the cutting edge. For bringing up the cuttings a certain amount of force is necessary, which is compounded of the peripheral force and the axial pressure of the drill. If the drill pressure P is too small, the drill is apt to

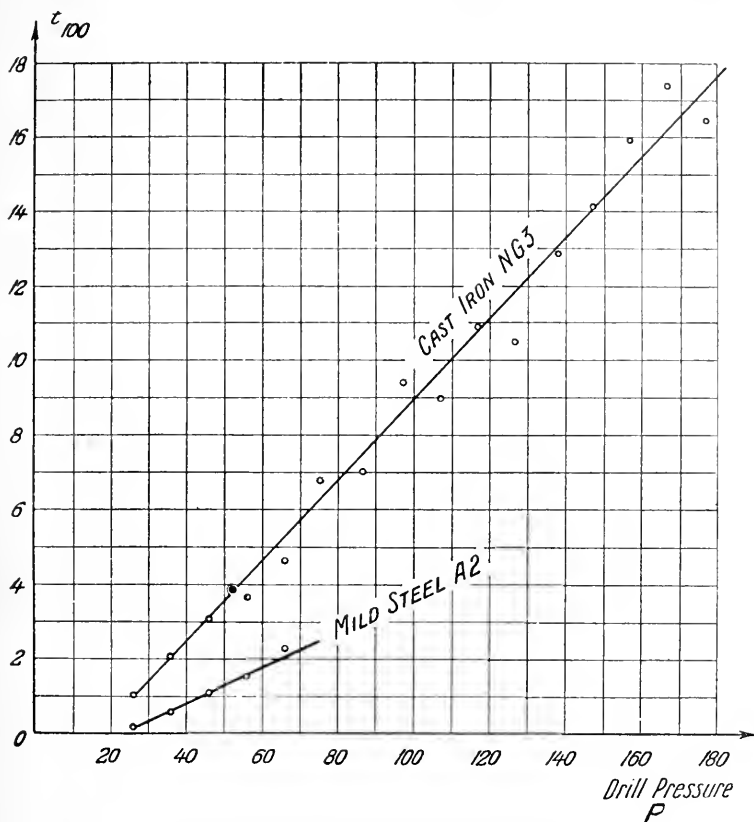


FIG. 20.—Influence of Drill Pressure P on Value of t_{100} for Cast Iron and Mild Steel.

be lifted by the cuttings accumulating and jamming underneath, which produces a break in the curve. The most uniform values were obtained with drill pressures of 56 to 75 kilogrammes, and this was in general the pressure used in the subsequent tests.

STANDARD MATERIAL FOR COMPARISON.

As the drill test can never yield absolute results, but only comparative values, which must of necessity be obtained under equal external conditions, the author made a very large number of tests with the object of discovering a suitable material to serve as a standard of comparison to which other materials could be referred. Such a standard material must fulfil two important conditions:—

1. It must admit of having performed upon it a very large number of tests without exhibiting any appreciable variation as regards machining properties.
2. It must be easily producible anywhere under identical conditions as regards chemical composition, and although of different origin, it must, subject to similar treatment, possess approximately similar machining properties.

The drill test should then proceed as follows:—

1. Drilling of the standard material.
2. Drilling of the material to be tested and compared.
3. Drilling of the standard material.
4. The mean values of tests (1) and (3) are taken as equal to 100, that is, the machining properties of the standard material equals 100, this value being denoted by B_N .
5. The mean values of tests (2) are reduced to terms in proportion to B_N , and this result gives the required relative machining properties B . For example, let the mean value of the standard material as found by tests (1) and (3) be t_{100} (depth of hole in millimetres obtained after 100 revolutions) = 5.2 millimetres, and the mean value of the material to be tested be $t_{100} = 3.1$ millimetres.

Then—

$$\frac{100}{5.2} = \frac{B}{3.1}$$

Therefore the required relative machining property of the test material is

$$B = 59.6.$$

As quoted above, Leyde¹ and Reininger² used cast iron as a standard material, a large stock of which was cast chiefly in the form of flat plates. Special importance was attached to getting plates of equal machining properties and hardness, which is only possible if the chemical composition and rate of cooling of the cast iron are the same. The author alludes further on to his experiments on the influence of rate of cooling on ball hardness and machining properties; as regards the abandonment of the drill test in many foundries, Professor Heyn has expressed the opinion that the reason lies in the difficulty of providing a large stock of standard pieces of uniform composition and properties for purposes of comparison.

The author has been at the pains of testing various kinds of cast iron under different external conditions as regards form of drill, pressure, speed of cutting, &c., and has made altogether 750 tests. From these tests he has concluded that as a standard of comparison cast iron is the most unsuitable material imaginable. Further attempts to find a mild steel that would serve as a standard also failed to lead to any satisfactory result. Finally, the author tested a variety of copper alloys, making 1035 tests of these altogether, but still without finding a material which satisfied the conditions. It almost appeared as if copper itself would prove suitable, but further tests with various coppers failed to substantiate the first impression. Delta metal at first gave very uniform values, but a second consignment of that material failed altogether. The only material which in some degree fulfilled the requirements as a standard for comparison was tombak, an alloy consisting of the two metals zinc and copper alone, which can be obtained fairly pure and can easily be mixed together. It would, of course, be necessary to prescribe carefully the several stages of manufacture of this standard material from the casting to the finishing of the bars, but that would not present any great difficulty.

After making upwards of 2000 tests, the author was driven to the conclusion that it is practically impossible for one person experimenting in a single laboratory to discover experimentally

¹ *Zeitschrift des Vereins Deutscher Ingenieure*, 1904, p. 169.

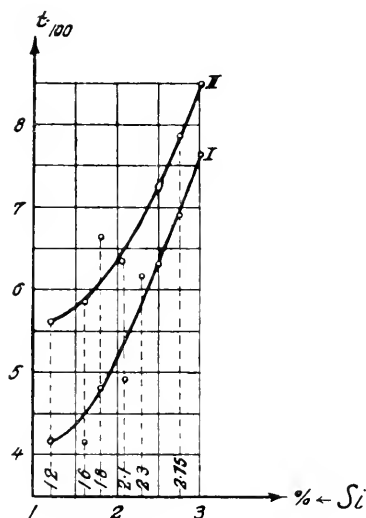
² *Giesserei Zeitung*, 1904, pp. 217 and 627.

a standard material for comparison, and that only an exchange of experience on the part of several independent investigators will lead to the desired result.

The lack of a recognised standard material, however, does not prevent the useful application of the drill test at any particular place, so long as the conditions laid down on pp. 6, 22, and 30 are observed. Thus the tests described at the end of the paper represent a connected series of investigations which can be performed without a standard material for comparison.

BALL-PRESSURE HARDNESS AND MACHINING CAPACITY OF BRASS WITH INCREASING PERCENTAGES OF LEAD.

The author's first experiments were directed to the testing of the machining properties of brass test-pieces with increasing



- I. Twist drill 10.5 millimetres diameter; revolutions per minute=197; P=106 kilog.
 II. Flat „ 8.0 „ „ „ „ „ „ =228; P= 75 „

FIG. 21.—Influence of Silicon Percentage on 5 Cast Iron Test-pieces of equal Cross Section. Mean of every 5 tests (50 tests in all).

lead percentages. The specimens were compounded and melted by the author. Twist drills 10.5 millimetres diameter of Novo

steel were used with a pressure P on the drill = 75 kilogrammes, and 197 revolutions per minute. Fig. 28 shows that t_{100} rises sharply, with an addition of $\frac{1}{2}$ to 1 per cent. of lead. It continued to rise at a relatively slower rate as the proportion of lead was increased, and at 12 per cent. of lead it reached a value of $t_{100} = 42$ millimetres. By the addition of 12 per cent. of lead the machining properties of the material was thus increased fourteen times as compared with brass containing no lead. One would not have expected this result, judging from the recorded ball-pressure hardness which reaches a maximum at 2 per cent. of lead and then falls again. The two curves afford a striking proof that the machining properties cannot be deduced from the ball-pressure hardness.

The influence of the silicon content on the machining properties of cast iron was ascertained in eight tests as follows: The test-pieces were of the dimensions shown in Fig. 22, Plate III., and in casting the underside of the mould was formed by a chill, the three other sides consisting of greensand. The fracture shows the chilling effect, which is also distinctly visible in the photograph. The chilled layer is thinnest in the test-piece with the highest silicon percentage, and gradually increases in thickness until it attains a depth equal to one-quarter of the whole section in the specimen lowest in silicon.

The holes from which Curve I. is plotted were drilled in the side 105 millimetres above the chilled underside, and those for Curve II. 90 millimetres above it. The tests were performed on the hardness drilling machine, Pattern I., as follows:—

Curve II.—Flat drill 8 millimetres diameter of high speed steel running at 228 revolutions per minute. Pressure P on the drill = 75 kilogrammes.

Curve I.—Twist drill 10.55 millimetres diameter of high speed steel running at 197 revolutions. $P = 106$ kilogrammes.

Both curves show a considerable increase in the machining properties as the silicon percentage rises. Unfortunately, for lack of time, the author was unable to make ball-pressure tests on these cast iron specimens, but hopes to do so at a subsequent date.

The superiority of the flat drill over the twist drill is plainly brought out by these tests, and is the more noticeable seeing that the pressure on the flat drill was less than that on the twist drill.

The influence of rate of cooling on the machining properties of cast iron was determined by means of the following tests. Two castings, in the form of a number of bars (Fig. 23), which had been formerly prepared for Professor Heyn for metallographical investigation were used, and the individual bars were drilled in the centre of their section and at their

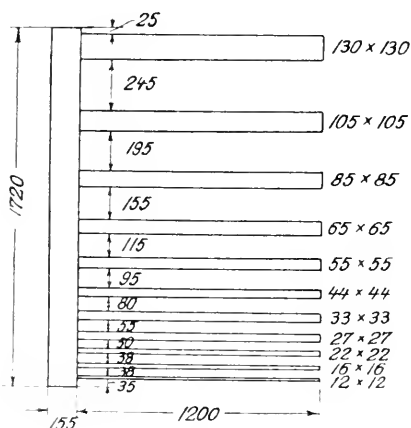


FIG. 23.

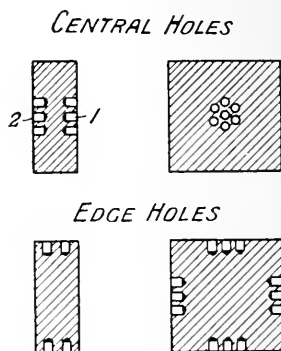


FIG. 24.

edges, as shown in Fig. 24. The two first holes were drilled exactly on the centre line of the section and the others in a circle round it, or as far as this was possible, in the bars of smaller sections. The holes at the edges were drilled in from the sides, and in these experiments, which were the very first tests made by the author, nothing could be detected which might indicate which had been the upper, lower, or vertical side in casting. A twist drill of high speed steel was used 8 millimetres diameter, running at 228 revolutions and loaded to 75 kilogrammes. After each hole the drill was re-ground on a special grinding tool for twist drills. The first casting, on which tests Nos. 11 to 21 were made, was a harder kind

of iron, and the second casting, on which tests Nos. 23 to 33 were made, was a softer iron.

The results of these tests are represented graphically in Fig. 25; the edges of the cross-section are plotted as abscissæ and the values of t_{100} as ordinates. The values of the holes in the edge are the mean of sixteen tests, those at the centre the mean of eight tests. The curves show plainly that the specimens M. 23 to 33, cut from the softer material, have

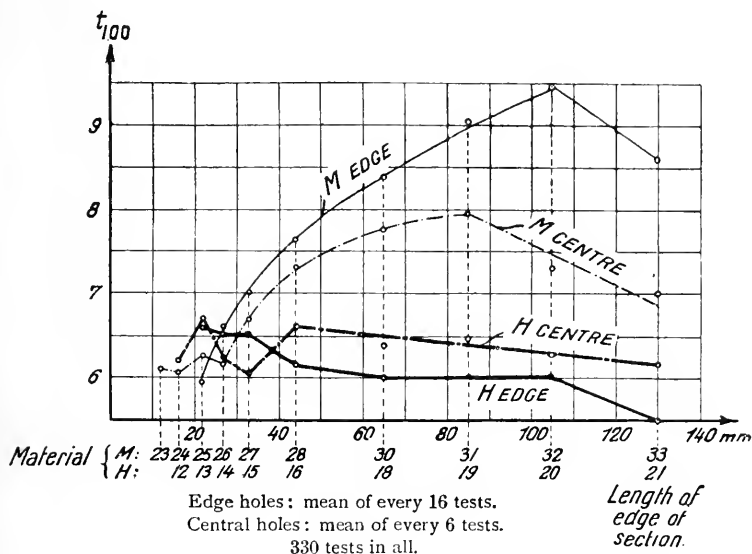


FIG. 25.—Influence of Rate of Cooling on Value of t_{100} for Cast-iron Bars cast from the same Ladle.

much better machining properties than specimens H. 11 to 21 from the harder material. Most striking is the comparison between the specimens M. 32 and H. 20 of equal section. The holes in the edge of specimen M. 32 show $t_{100} = 9.45$ millimetres, and in specimen H. 20 $t_{100} = 6$ millimetres. Another curious feature is that in the specimens H. 11 to 21 the holes at the side show lower machining properties than those in the centre, whereas in specimens M. 23 to 33 the conditions are exactly reversed. The differences in the machining properties between the central and the side holes are greatest in specimen

M. 32, the side holes giving $t_{100} = 9.45$ and the central holes $t_{100} = 7.3$, or 77 per cent. of the first value of the former figure. Against some of the specimens the sulphur and phosphorus is noted (Fig. 26), and it will be observed that the parts at the edge are richer in phosphorus and sulphur than the central portion, but in what degree phosphorus and sulphur affect the machining properties, other conditions being equal, could not be determined from these few experiments. This should prove a useful field of investigation for the future.

It is of special interest in the case of specimens M. 23 to 33 to tabulate the distribution of graphite, bending resistance and machining properties (Fig. 26), since the quantity and the mode of distribution of the graphite have a determining influence on the physical properties of cast iron. The graphite content and the number of graphite flakes per square millimetre of the surface were determined separately. The average composition of the cast iron, according to Heyn, was:—

	Per Cent.
Total carbon	3.38
Silicon	2.51
Manganese	0.81
Phosphorus	0.56
Sulphur	0.095

The graphite content was lowest in the centre of the thinnest bars, and increased regularly as the section of the bars increased, reaching a maximum in specimen M. 30 with section 65×65 millimetres, which was maintained in the specimens of larger section. The machining properties also rose considerably, but reached their maximum of $t_{100} = 7.95$ in the central holes of specimen M. 31, 85×85 millimetres square, and then fell. In the outer zones the maximum graphite content is first reached in specimen M. 31, 85×85 millimetres square, but the maximum machining properties of the side holes is first reached in specimen M. 32, 105×105 millimetres square. It is of interest to note that before these tests were made, O. Leyde, who for many years was manager of the foundry of Ludwig Loewe, in Berlin, had recognised that specimen M. 31 showed the most suitable grain for practical purposes. Finally, the curves of Fig. 26 show that the

bending resistance curves follow a course similar to that recording the number of graphite flakes, an agreement which Heyn¹ has already pointed out.

The machining properties t_{100} and the ball-pressure hardness P 0.05 of these specimens are summarised in Fig. 27. The ball-pressure hardnesses were obtained with the Martens-

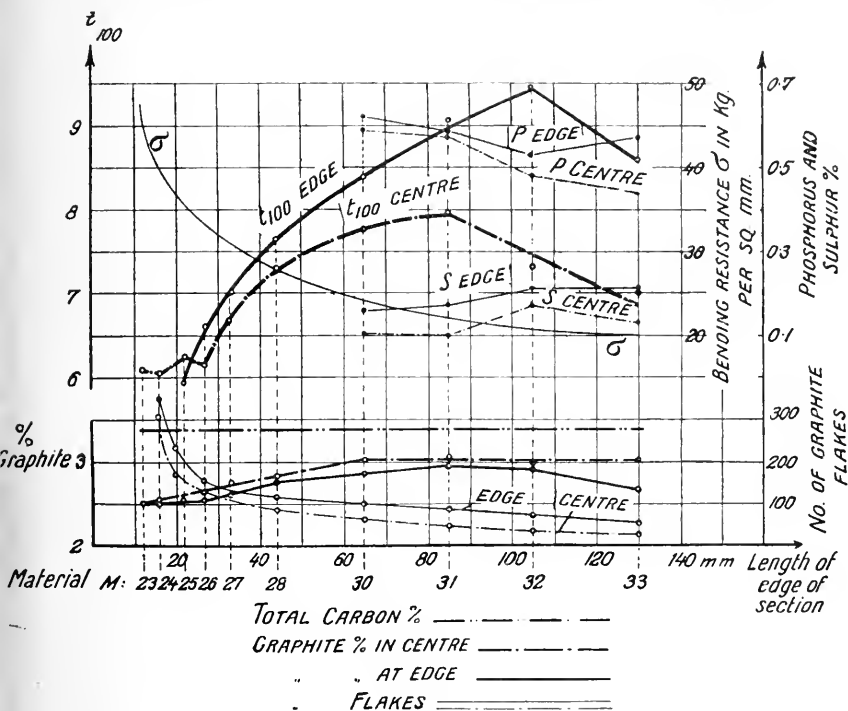


FIG. 26.

Heyn device.² The ball-pressure hardness P 0.05 is the pressure in kilogrammes required to impress a steel ball 5 millimetres diameter to a depth of 0.05 millimetre. The tests again show clearly that the machining properties t_{100} and the ball-pressure hardness P 0.05 are not comparable, and that it is therefore absolutely necessary that the testing

¹ *Stahl und Eisen*, 1906, p. 1295.

² A. Martens and E. Heyn, *Zeitschrift des Vereins Deutscher Ingenieure*, 1908, p. 1719.

of the machining properties should be assigned its proper importance in the regular testing of materials.

The relations between tenacity, ball-pressure hardness, and machining properties of some mild steels are represented in curves in Fig. 30, which show the average values of 420 tests. For these tests flat drills of 8 millimetres diameter,

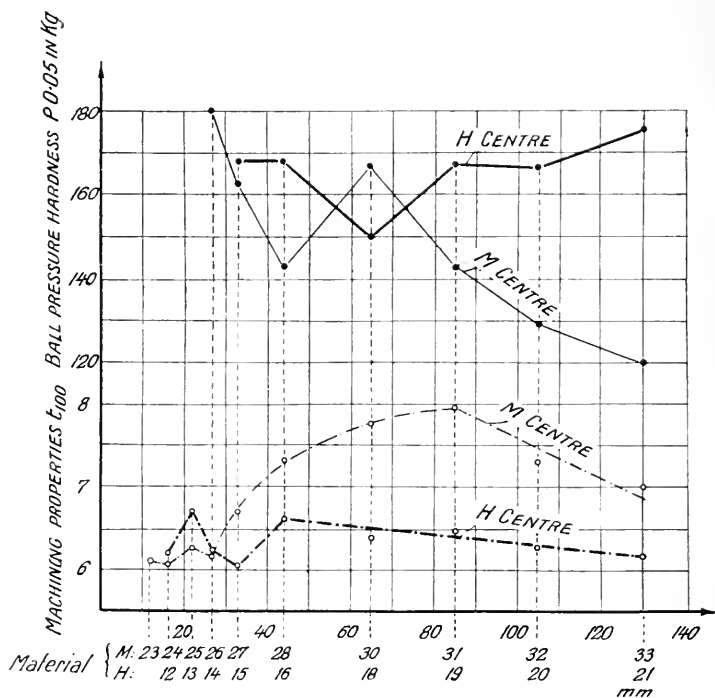


FIG. 27.—Influence of Rate of Cooling on Ball-pressure Hardness P 0.05 and Machining Capacity t_{100} of various Cast-iron Test-pieces. P 0.05 are mean values of every 10 tests. 130 tests in all.

Length of
edge of
sector

with angle ϕ of the point = 130° , and clearance angle $\alpha = 20^\circ$, according to Fig. 16, were employed, the drill pressure $P = 76$ kilogrammes, and the drill running at 710 revolutions per minute.

In general, the machining properties improved as the tensile strength rose. It is noteworthy that materials A1, A2, and A3 show machining properties of only about 70 to 86

per cent. of that of materials B01, B03, and B05, with approximately equal tensile strength. The ball-pressure hardnesses $P\ 0\cdot05$, recorded in the same figure, also show an increase with increasing tenacity, a fact which is well known, but they cannot be brought into comparison with the machining properties.

The values of the machining properties of several typical materials for turret-lathe work, which are noted to the right of Fig. 30, show higher results with a lower ball-pressure hardness,

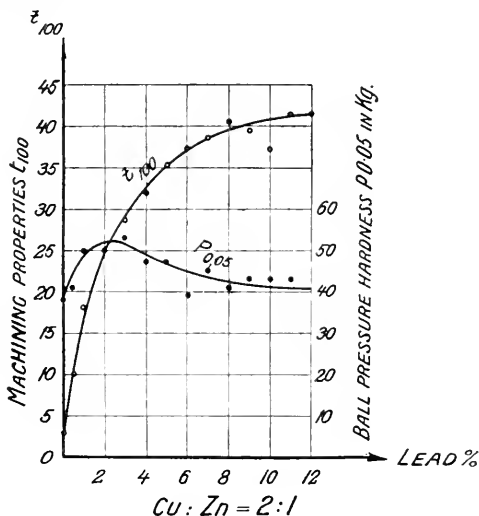


FIG. 28.—Influence of Lead Percentage on Ball-pressure Hardness $P\ 0\cdot05$, and Machining Capacity t_{100} of Cast Brass.

$P\ 0\cdot05$, than the above-mentioned kinds of mild steel. Unfortunately, the author had not time to investigate the reasons for this, which presumably are to be sought for in the chemical composition. Further, the influence of lubricating the drill with oil, in the case of material BRE 1, was investigated, and it was found that the value t_{100} rises from 3.1 to 5.4. The separate curves for the oil lubricated material certainly show greater variations than those for the holes which were drilled dry, an occurrence already frequently observed by the author. For testing the machining properties of mild steel it would appear advisable to drill holes both

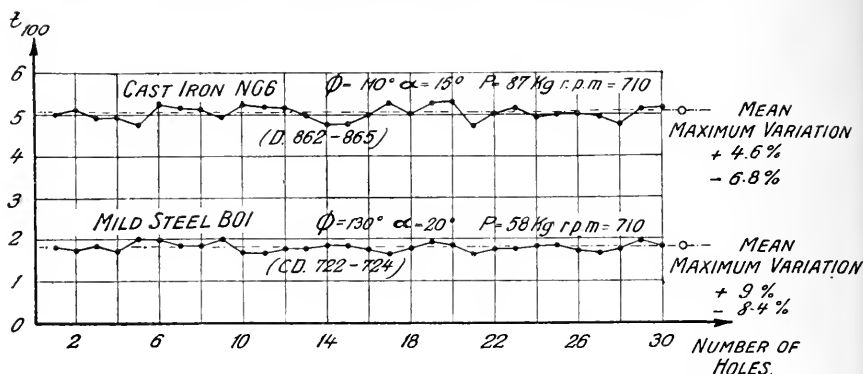


FIG. 29.—Variations of t_{100} in each 30 Tests. Drill re-ground for first hole in each case.

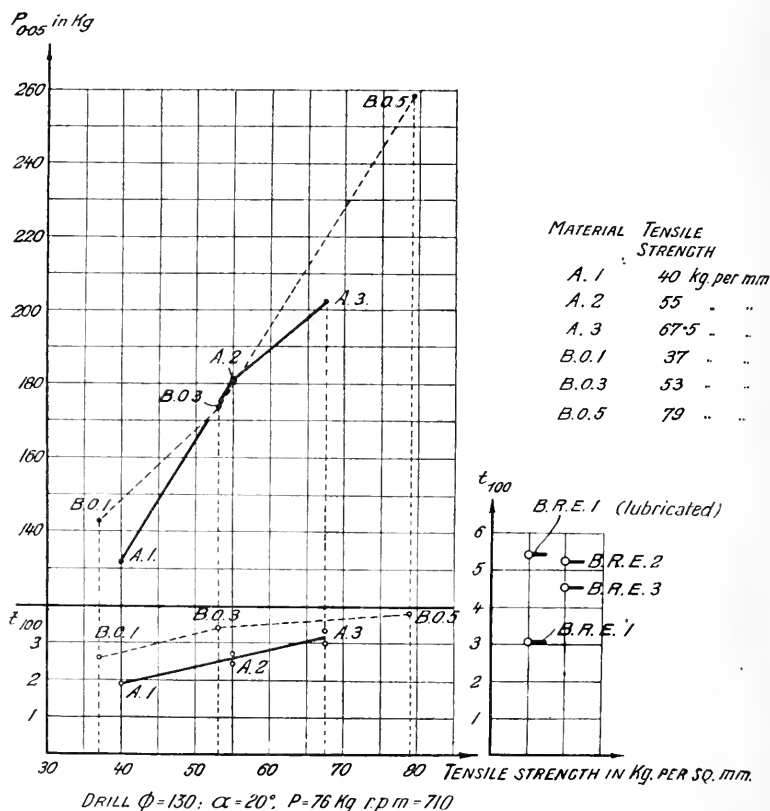


FIG. 30.—Relations between Tensile Strength, Machining Capacity t_{100} , and Ball-pressure Hardness $P 0.05$ of Mild Steel.

with and without lubrication, a question which the author proposes shortly to investigate.

HARDNESS DRILLING MACHINE, PATTERN III.

(*Kessner Type*).

The defects which manifested themselves in the hardness-testing drilling machines, Patterns I. and II., particularly the imperfect engagement of a small pinion in a toothed rack, led to the development by the author of Pattern III., in which the use of the rack is entirely dispensed with. A close examination of the new arrangement described below proved that the total friction within the machine was considerably reduced, and was of a much more uniform degree. In consequence, the variations in the test results were very much less.

In Pattern III. of the hardness-test drill the indicator arrangement was the same as in Pattern II. (Fig. 32, Plate IV.). The feed-wheel, 14*a*, with the small pinion 16, was entirely removed. The feed of the drill spindle 1—the effective pressure on the drill—is actuated (Fig. 31) by the lever 13, which swings on the fixed pivot 32, exerts pressure on the drill spindle, and carries the weight *p* on the segment 14. The fixed bracket 33*a* was at first formed of two plates firmly bolted to the lower guide block of the drill spindle. It would, however, be better in reconstructing the machine to cast a solid bracket on the lower guide bracket of the drill spindle instead of using the two plates.

The forked lever 13 bears down on a collar 29, provided with two gudgeons, the collar pressing direct on the ball-bearing of the drill spindle. The theoretical drill pressure P_0 is always the same whatever the position of the forked lever, namely:—

$$P_0 = \frac{p \times b}{a}$$

The semicircular notches in the forked ends of the lever in which the gudgeons rest have a slightly larger diameter than the gudgeons themselves, in order to allow for the

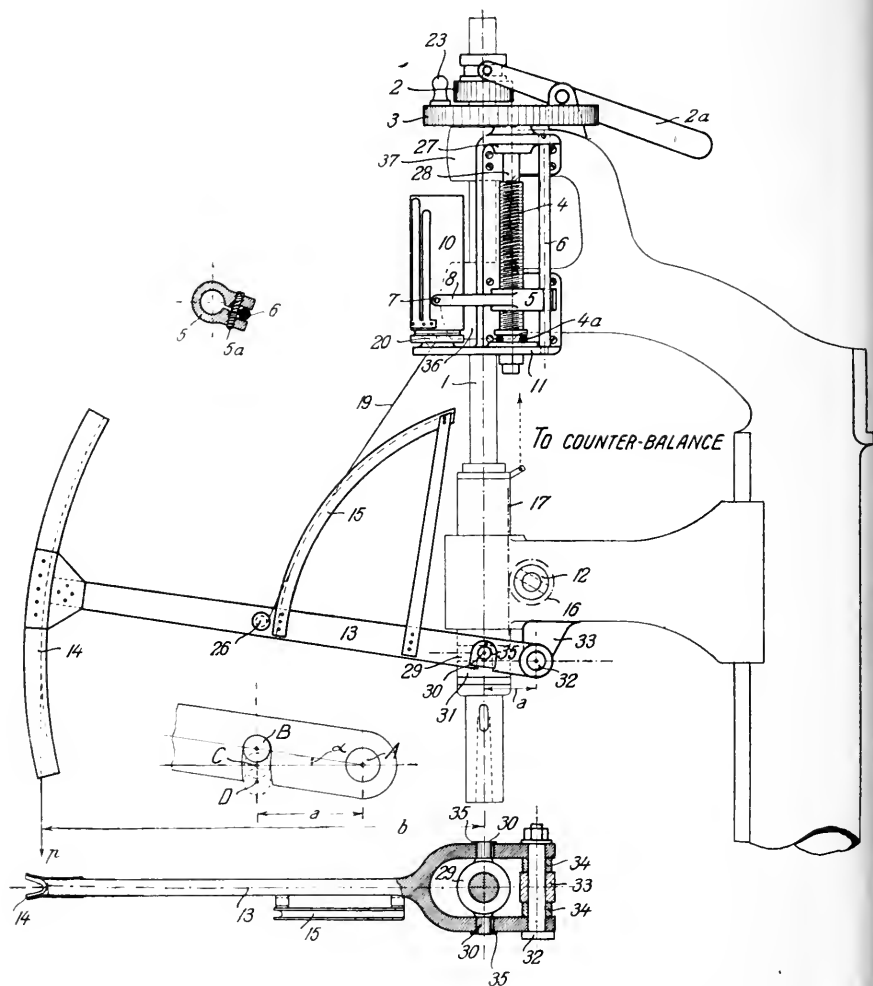


FIG. 31.—Apparatus for Testing Machining Properties of Materials (Hardness-testing Drill). Pattern III. (Kessner Type.)

necessary play required on account of the radial path of the lever end. But the radial movement is too small to affect

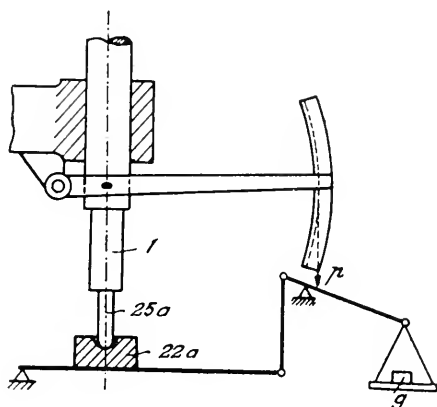


FIG. 33.—Determination of the actual Drill Pressure P with the Hardness-testing Drill Pattern III.

the drill pressure P_0 and the accuracy of the results. The depth of the deepest hole bored under an unusually heavy pressure was 11 millimetres, and in this case :—

$$\begin{aligned} BD &= 11 \text{ millimetres.} \\ \frac{1}{2}BD &= BC = AC \times \tan \alpha. \\ 5.5 &= 40 \times \tan \alpha. \\ \alpha &= 7^\circ 49' 30''. \end{aligned}$$

Putting this at the round value of 8° , the greatest displacement of the gudgeons in the lever fork will amount to :—

$$BA - CA = \frac{CA}{\cos \alpha} - CA = 40 \left\{ \frac{1}{\cos \alpha} - 1 \right\} = 0.4 \text{ millimetres.}$$

The notches in which the gudgeons rest should therefore have a radius of 0.2 millimetre larger than that of the gudgeon. A further advantage of this arrangement is that by changing the segment 15 the scale of depths of hole can be enlarged on the diagram to any desired size.

In determining the actual pressure P on the drill, the same procedure was followed as that described on page 24. With the hardness-testing drilling machine, Pattern III., the theoretical drill pressure $P_0 = 17.1 p$ (p being the weight in

kilogrammes suspended on the lever). The curves for P and P_0 follow a parallel course (Fig. 34), proving that the total friction within the machine does not change with an increasing load. From the commencement the curve for P lies above that for P_0 , because the weight of the lever is purposely left unbalanced, thus adding pressure on the drill.

The hardness-testing drilling machine, Pattern III., fulfils

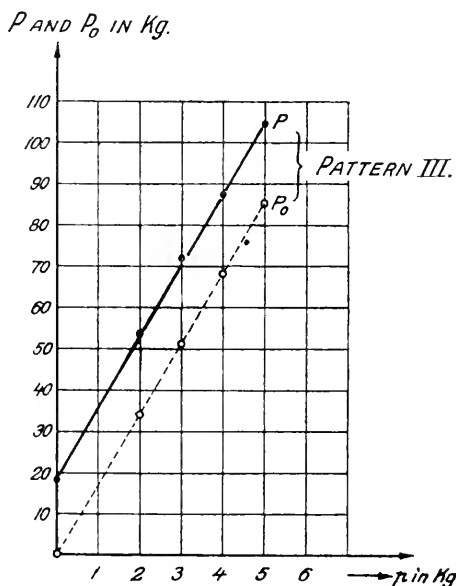


FIG. 34.

the conditions laid down by the author, and has given very regular results in operation.

The testing of tool steels by means of the hardness-testing drilling machine was the latest endeavour of the author. It was intended to make flat drills, as shown in Fig. 16, from various kinds of steel, and to drill hard material as far as practicable. It is to be assumed that the drill will become blunt, which will be indicated on the drilling curves. For lack of time the author has unfortunately not been able to complete these tests, but proposes now to proceed with this work.

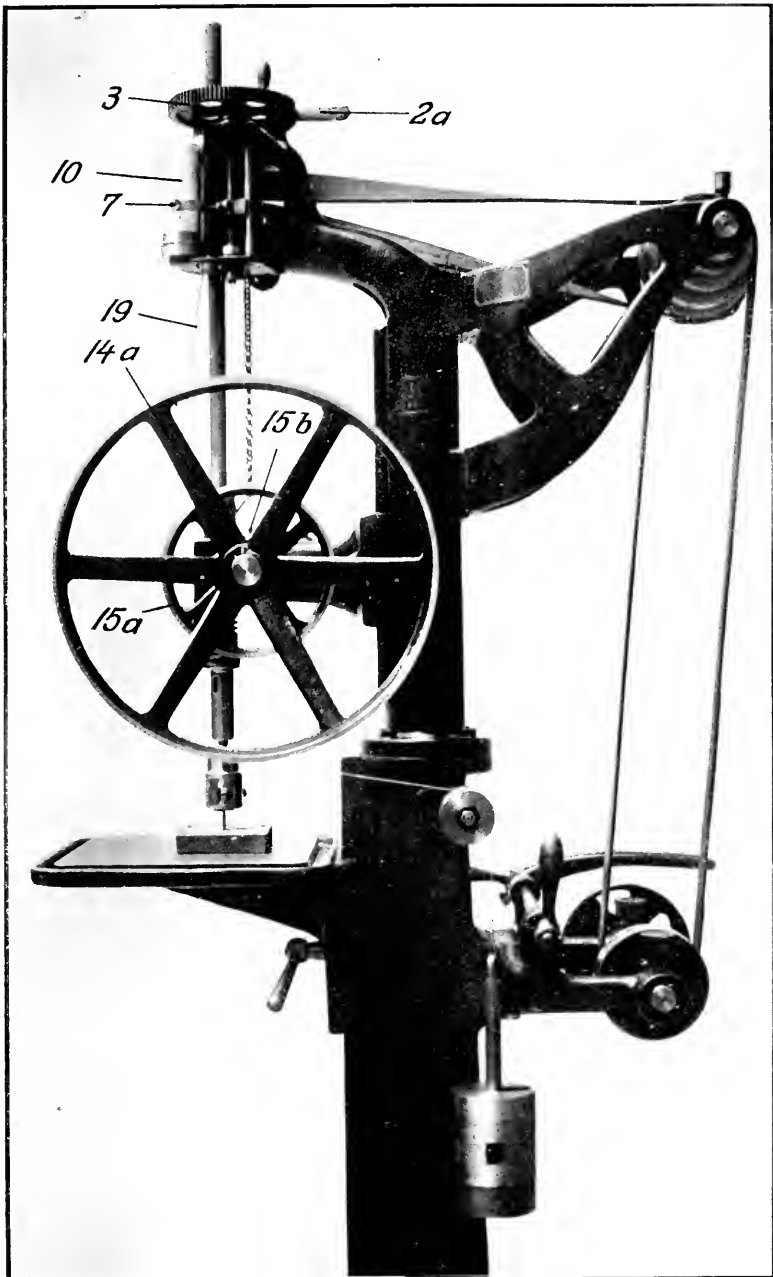


FIG. 9.—Hardness Testing Drill. Pattern II. (Kessner Type.)





$$\begin{aligned}\phi &= 110^\circ \\ \alpha &= 30^\circ\end{aligned}$$

$$\begin{aligned}\phi &= 110^\circ \\ \alpha &= 20^\circ\end{aligned}$$

$$\begin{aligned}\phi &= 110^\circ \\ \alpha &= 10^\circ\end{aligned}$$

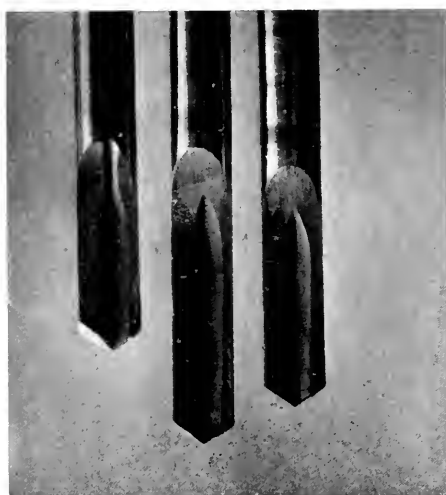


FIG. 15.—Form of Drill finally adopted.



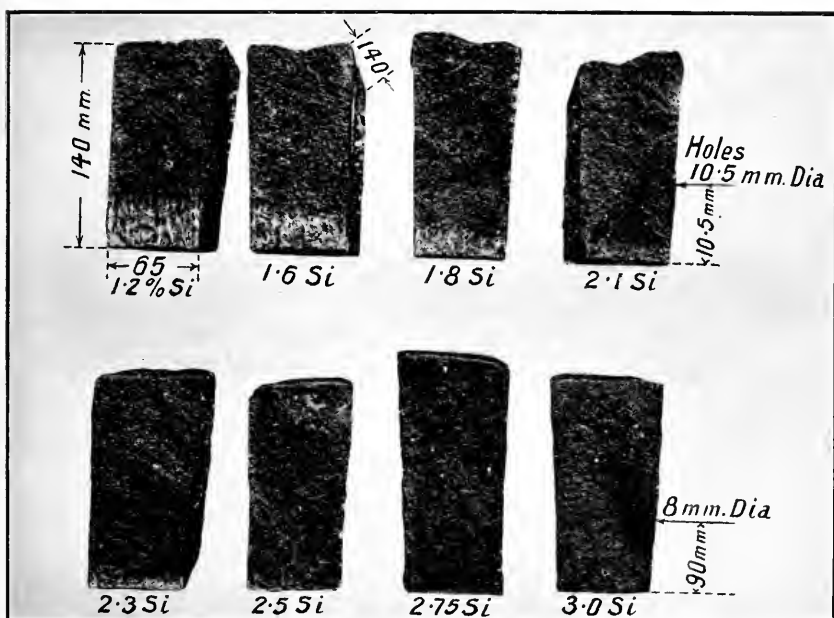


FIG. 22.

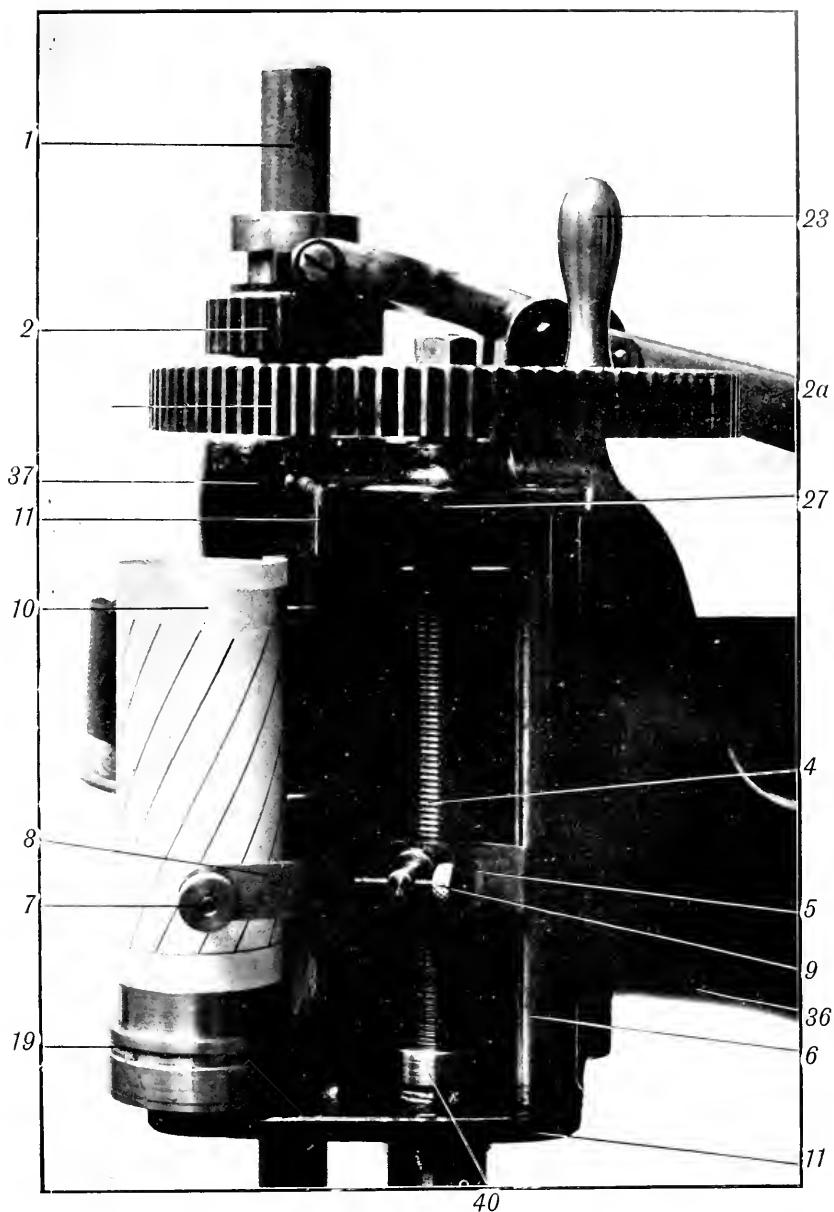


FIG. 32.—Indicator of the Hardness-testing Drill. Patterns II. and III.



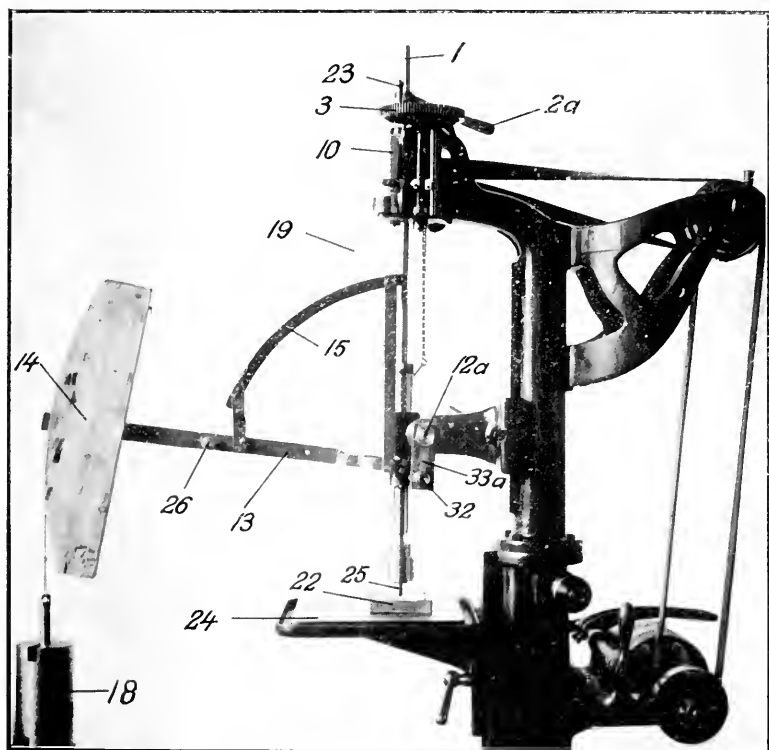


FIG. 35.—Hardness-testing Drill. Pattern III. (Kessner Type.)



SUMMARY.

The results of the present work may be shortly summarised as follows :—

1. Of three different patterns of hardness-testing drilling machines which have been designed, constructed, and tested, the last one, Pattern III., has fulfilled the required conditions and has proved satisfactory as an appliance for testing the machining properties of metals.
2. For hardness drilling tests there has been found a suitable form of drill with which about 2000 tests have been performed. At the same time, the influence of the cutting angles of the drill on the depth of hole attained has been determined.
3. Evidence has been afforded that it is impossible from the ball-pressure hardness of a material to draw any conclusion as to machining properties.
4. The influence of adding lead in varying proportions to brass upon the machining properties of the brass has been ascertained.
5. The influence of the rate of cooling on the machining properties of cast-iron bars of the same quality of iron has been determined.
6. The influence of silicon on the machining properties of cast-iron test-pieces of equal cross-section has been determined.
7. Some relations between tenacity, ball-pressure hardness, and machining properties of various kinds of mild steel have been ascertained.

In the performance of this research work the author has carried out 3526 drill tests and 750 ball-pressure tests, all of which were performed in the Mechanical Technological Laboratory of the Royal Technical High School at Gross-Lichterfelde. By the kind permission of Professor Heyn, Chief of the Laboratory, all the laboratory appliances were placed at his disposal, and the author wishes most especially to present his acknowledgments to Professor Heyn for the facilities afforded him for carrying out the work.

List of Materials Tested.

No.	Material.	Remarks.
		Per Cent. Per Cent. Per Cent.
		Zinc, 33·33; copper, 66·60; Lead 3
Mi.	Cast brass.	
M. 21	"	" " " 1·0
M. 3	"	" " " 2·03
M. 4	"	" " " 0·505
M. 5	"	" " " 4·0
M. 6	"	" " " 5·05
M. 7	"	" " " 1·0
M. 8	"	" " " 6·0
M. 9	"	" " " 7·0
M. 10	"	" " " 8·0
M. 11	"	" " " 9·0
M. 12	"	" " " 10·0
M. 13	"	" " " 4·0
M. 14	"	" " " 8·0
M. 15	"	" " " 11·0
M. 16	"	" " " 12·0
M. 17	"	12 mm. round bar revolver material.
M. 18	"	28·5 " "
M. L	Durana metal.	20 " "
22 B	" brass.	30 " "
93/7	" phosphor bronze.	
M. M. B	"	
M. B. VII.	" manganese bronze.	25 " " 16 mm. square bar.
M. R. D. 1	Brass.	25 " "
M. R. H. 1	"	30 " "
M. R. F. 1	"	28·5 " "
M. R. E. 2	"	29·0 " "
K. 2	Drawn copper.	12 " square bar.
T. 2	Tombak.	12 " "
R. 2	Reinicka.	12 " "
M. 19	Brass.	12 " "
D. 2	Delta metal.	14 " "
K. 3	Copper.	20 " "
K. 4	Drawn copper.	15 " "
M. 27	" brass.	30 " round bar.
T. 3	Tombak.	12 " square bar.
G. 22	Cast iron.	Cylinder metal, first tapping.
G. 23	" "	" " second tapping.
G. 24	" "	Engine metal, first tapping.
G. 25	" "	" " second tapping.
N. G. 1-6	Standard cast iron.	
B. O. 1	Mild steel.	10 millimetres square bars.
B. O. 3	" "	Square bars; elastic limit 37 kg. per mm ² , elongation 25 per cent.
B. O. 5	" "	Square bars; elastic limit 53 kg. per mm ² , elongation 21 per cent.
B. R. E. 1	Soft steel for bicycle wheel hubs.	
B. R. E. 2	Pedal steel.	
B. R. E. 3	Crank spindle steel.	
A. 1	Mild steel.	40 kg. per mm ² .
A. 2	" "	53 " "
A. 3	" "	65 " "

AN INVESTIGATION OF LIQUID CONTRACTION IN CAST IRON.¹

BY GEORGE HAILSTONE (BIRMINGHAM).

INTRODUCTION.

THE object of this investigation was to solve if possible the reason why certain grey iron castings used in the engineering world contain subcutaneous cavities, which are a source of great trouble to high speed and hydraulic engineers, inasmuch as very frequently a casting when machined is apparently sound, but when subjected to a hydraulic test is found to be leaky.

“Liquid contraction” is a term originated by Longmuir to distinguish the local contraction which takes place in the heavy part of a casting when the outside skin of that casting has solidified.

The terms “sinking,” “draw holes,” and “sponginess” are not dealt with in this memoir.

“Solid contraction” is the natural contraction which metals undergo when passing from the hot to the cold condition. Solid contraction is also known as “shrinkage.”

As an introduction it will be as well to consider the real difference between solid and liquid contraction, and also to review some of the conditions which prevail, and which must be taken into account in dealing with this most difficult and complex subject.

There is but little doubt that these two forms of contraction are closely associated with each other as regards occurrence and extent; that is, the conditions which conduce to a minimum liquid contraction are those which conduce to a minimum solid contraction or shrinkage. On the other hand, the conditions which conduce to a maximum liquid contraction are also those which produce a maximum solid contraction.

¹ Received February 28, 1913.

If, for example, a 5-cwt. ladle full of ordinary No. 3 foundry iron be taken and run into a bar 12 feet long, the solid contraction or shrinkage would be about one-eighth of an inch per foot of the length. Supposing the same iron is run into a cube, the solid contraction would be much less than in the previous case, but the liquid contraction will be found to be greater if the cube be cut through the centre, and owing to the drawing away of the iron to the exterior, which was the last to solidify.

The lighter part of any one casting can draw liquid iron from a contiguous heavy part, and thus lessen its own solid contraction. Each part, in its turn, takes from the more fluid iron next to it, and the part which has molten iron taken from it, and cannot get a fresh supply, exhibits what we term "shrink-holes" or "sinking," the condition of which are different from "liquid contraction."

Most foundrymen agree that castings run with "dull" or "cold" iron do not so readily "sink" or "draw" as when the metal is run "hot." This may be so, and is certainly true when the casting is only examined for surface defects.

An endeavour will be made later to prove that whilst "sinking" and "drawing" in a casting may be reduced to a minimum by pouring with "dull" iron, liquid contraction is reduced to a minimum by pouring with iron as hot as it is possible to obtain in ordinary cupola furnace melting, without the metal being actually on the boil.

The subject of liquid contraction in cast iron was first brought to the author's notice by Mr. F. J. Cook (Past-President of the British Foundrymen's Association) some years ago, to whom the author offers his best thanks for the kindness shown in providing most of the samples that have been used in this investigation.

In a certain important class of castings which have now been made for many years, and with which the author is familiar, one part of the section is in the shape of the letter "K," and this particular spot at intervals used to give a great amount of trouble, being leaky when the casting was subjected to a hydraulic pressure test. No indication of the cause was given by chemical analysis, so an investigation was commenced

by casting "K" bars from each of the pig irons being used, of sectional dimensions coinciding roughly with the thickness of the castings being made. It was then found that one of the irons (a so-called refined iron) was very prone to liquid contraction. This led to the discarding of this particular brand of iron, and consequently to a diminution of the trouble. As a further preventive, a system was inaugurated of casting each day a set of these test bars so as to give at once an indication of any recurrence of the trouble.

That this trouble cannot be detected by chemical analysis seems to be proved by the following tests on two irons:—

Analysis.

	No. 1.	No. 2.
	Per Cent.	Per Cent.
Graphitic carbon	2·280	2·295
Combined carbon	0·720	0·736
Silicon	1·306	1·430
Sulphur	0·056	0·064
Phosphorus	0·465	0·492
Manganese	0·432	0·432
Iron by difference	95·741	94·551

Mechanical and Physical Tests.

	No. 1.	No. 2.
Shrinkage bar, No. 1	0·203 inch	0·209 inch
" No. 2	0·206 inch	0·209 inch
Transverse test bar, $\frac{1}{2}$ -inch, No. 1, 12-inch centres	480 lbs.	470 lbs.
" " No. 2, " " "	485 lbs.	465 lbs.
Deflection, $\frac{1}{2}$ -inch bar, No. 1	0·150 inch	0·156 inch
" " No. 2	0·150 inch	0·156 inch
Transverse test, 1-inch bar, No. 1, 12-inch centres	3,696 lbs.	2,856 lbs.
" " No. 2, " " "	3,584 lbs.	2,940 lbs.
Tensile test bar, No. 1	16-ton, sq. in.	11·1-ton, sq. in.
" " No. 2	15·5-ton, sq. in.	8·3-ton, sq. in.
Condition of "K" bar	Perfectly solid	Sunk & spongy

Practically, the chemical composition of each is identical, and might be expected to give similar physical and mechanical results, but actually the test results were widely different, and so prone was the one to the trouble mentioned, viz., "liquid contraction," that it also affected the centre of the

round bar cast 1.5-inch diameter for the tensile test, causing a particularly low result for this class of iron.

It will be noticed that the difference is not so well marked with the small half-inch square bars for the transverse test, owing to the quicker rate of cooling, but all the others are affected.

GENERAL SCHEME.

The material used in this investigation has been for the most part cast iron, melted in the cupola, for various classes

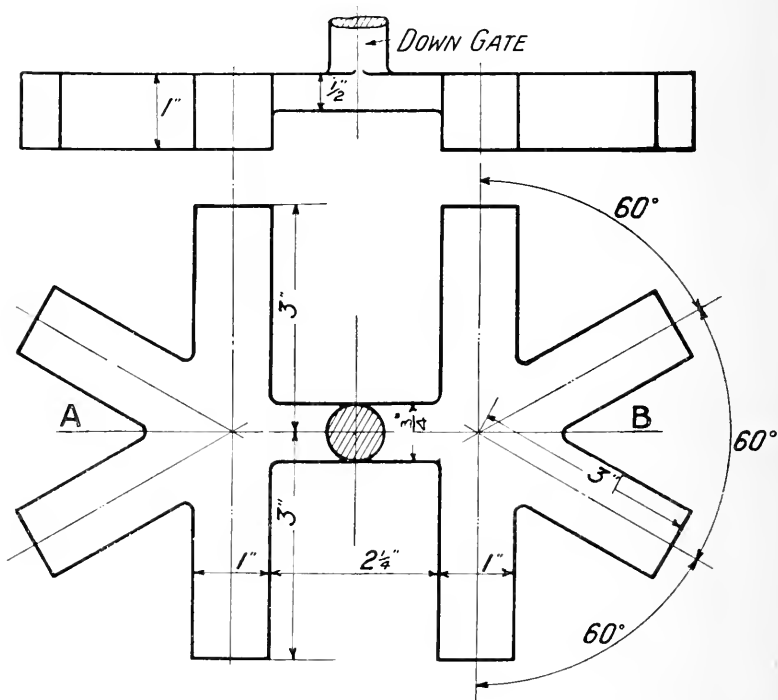


FIG. 1.

of castings, a ladle of any one mixture being poured at certain intervals into "K" test bars.

For the purpose of these tests two bars, each in the shape of the letter "K" of the dimensions shown in Fig. 1, were

investigation are not sound; in these cases, two sets of measurements were taken (1) from the affected areas, and

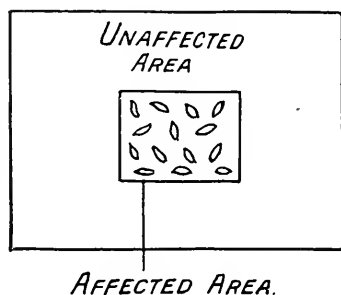


FIG. 2.

(2) from the unaffected areas, as shown diagrammatically in Fig. 2.

Twelve measurements of each constituent were taken at random from each area, and the mean is shown in the investigations that follow.

Specific Gravity of Samples.

A cube of one-half inch was machined from one of the short legs of the "K" bar from the same position in each case to determine the specific gravity of the sample.

Hardness of Samples.

The section of samples used for microscopical measurements were used for the hardness tests, the hardness tests being obtained by Shore's scleroscope, and the indentations being afterwards measured by the microscope micrometer. It was found, however, that the scleroscope did not give a constant rebound with a given indentation, therefore in the curves that follow the specific gravity or density of a specimen is plotted against the diameter of the indentation.

Twelve readings in each case were taken, the mean being quoted in the investigation.

Influence of Gases on Liquid Contraction.

The apparatus used was similar in design to that used by Carpenter and described by him in the *Journal of the Iron and Steel Institute*, 1909, No. II. p. 105.

The specimens used were turned out of one of the long legs of the "K" test bar from the same position in every case, the size being 2 inches long by 0.5 inch in diameter, weighing roughly 45 grammes.

The samples were heated at a temperature of 950°C . for twelve hours, and the results are quoted in cubic centimetres evolved per gramme of samples.

Methods of Chemical Analysis.

The drillings for the analysis of the samples were obtained by means of a half-inch twist drill from two holes right through the short leg of the "K" test bar and well mixing the whole together.

Total Carbon.—Dry combustion with copper oxide (Arnold and Ibbotson, p. 52).

Combined Carbon.—Difference between total and graphitic carbon.

Silicon and Sulphur.—2 grammes taken placed in No. 6 porcelain dish, 10 cubic centimetres concentrated HNO_3 added, heated until all action ceases, cooled 15 cubic centimetres HCl and 0.5 gramme KClO_3 added, take to dryness, baked for half-hour, and allowed to cool. Add 15 cubic centimetres HCl , take down almost to dryness, add 10 cubic centimetres HCl and 50 cubic centimetres hot water and filter. Wash with hot dilute HCl (5 per cent. solution) and finally with H_2O &c.

Filtrate.—Dilute to 250 cubic centimetres, raise to boiling, add 20 cubic centimetres BaCl_2 solution 10 per cent., boil for two minutes, allow to stand overnight, filter, wash well with cold dilute HCl , dry ignite, &c.

Phosphorus.—Molybdate estimation, 1 gramme taken.

Manganese.—Sodium bismuthate estimation, 1.1 gramme taken.

THE AUTHOR'S RESEARCH.

First Series.

Three sets of bars, A B and C, were cast from the same ladle of metal at intervals of time to ascertain the effect of casting temperature.

The bars were cast from cupola metal, the blast-pressure used being 12 oz. The appearance of the metal on being tapped from the cupola was that of good hot metal, the first of the bars being cast at this temperature.

The "K" bars, hardness, tensile, and transverse bars for each set were cast in one mould, from separate runners.

The hardness bars were cast $1\frac{1}{2}$ inch in diameter and tested by the drill method, *i.e.* the numerals quoted being the revolutions of a drill to attain a depth of 1 inch in the bar.

The tensile bars were cast $1\frac{1}{4}$ inch in diameter and turned down to $\frac{1}{2}$ square inch sectional area.

The transverse bars were cast $1\frac{1}{4}$ inch square, 14 inches long, and machined down to 1 inch square, and tested on 12-inch centres.

Keep's bar results are the mean of two cast in the same mould.

Fig. 3, Plate VI., shows the appearance of the "K" test bars as removed from the mould before removing the "gate."

The appearance of the fractured "K" test bars is shown in Fig. 4, Plate VII.

Bar A is a close-grained solid of a medium grey colour. Bar B shows a slight opening of the grain at the centre. Bar C shows the centre of the specimen to be porous, and large holes are to be seen towards the top side of the bar as cast.

The colour of the fractured bars in each case is that of a light grey, and the appearance of each of the "K" bars before being fractured was that of a sound casting.

The following table gives the analysis and specific gravity of the bars:—

TABLE I.

	Bar A.	Bar B.	Bar C.
Order of Casting.	First.	Second.	Third.
Constituents.	Per Cent.	Per Cent.	Per Cent.
Total carbon	3·089	3·118	3·162
Graphite carbon	2·569	2·609	2·657
Combined carbon	0·520	0·509	0·505
Silicon	1·477	1·479	1·497
Sulphur	0·117	0·120	0·086
Phosphorus	1·013	1·019	1·069
Manganese	0·368	0·337	0·312
Specific gravity of bars	7·283	7·228	7·130

From the analysis of the samples it is interesting to note that the higher the temperature at which the bar is cast the greater the percentage of combined carbon present. Also that the bars A and B contain more sulphur than bar C, although they are the more sound.

Table II. gives the results of the mechanical tests on the tensile, transverse, and hardness bars, cast in conjunction with the "K" bars.

TABLE II.

Bar.	Time Cast.	Interval off Time between Casting.	Keep's Test Bars.			Appearance of Fractured "K" Bar.	Hardness by Drill Method.	Tensile Strength.	Transverse Strength. Bar 1 inch sq. tested on 12 inch centres.
			Shrink-age.	Transverse Strength.	Deflection.				
	P.M.	Min.	Inch.	Lbs.	Inch.			Tons per Sq. In.	Cwt.
A	5·33	...	0·162	550	0·18	Solid	68·0	13·2	26·5
B	5·38	5	0·158	501	0·16	Practically solid	65·5	10·5	25·5
C	5·41	3	0·153	470	0·13	Blowholes	60·0	9·2	22·0

From the foregoing mechanical tests it will be seen that the strongest and best iron was obtained from the bars that were cast at the hottest temperature, and the weakest bars were those cast at the lowest temperature.

Measurement of Constituents.

On polishing the sections for the measurement of the microscopical constituents it was found that bars A and B exhibited an even surface, whilst bar C contained blowholes only, with no evidence of liquid contraction. Only one set of readings were taken in each case, and Table III. gives the mean measurements of twelve readings taken.

TABLE III.—*Size of Microscopical Constituents.*

FLAKES OF GRAPHITE.			
Bar.	Length, Inch.	Breadth, Inch.	Area, Square Inch.
A	0·00215	0·0000612	...
B	0·00347	0·0001003	...
C	0·00503	0·0001102	...
PLATES OF GRAPHITE.			
A	0·001205	0·000578	0·0000006965
B	0·00144	0·000823	0·000001185
C	0·00217	0·001558	0·00000338
MANGANESE SULPHIDE AREAS.			
A	0·000259	0·000254	0·00000006578
B	0·000195	0·000195	0·0000003802
C	0·000215	0·000205	0·0000004407

From the preceding table it will be seen that the hotter the iron when cast the smaller the flakes and plates of graphite, whilst in the case of bar A, which was the hottest cast, it will be observed the manganese sulphide areas are the largest, the two cases B and C being approximately the same size.

Hardness of Samples taken by Shore's Scleroscope.

Table IV. gives the hardness of the sections used for microscopical measurements, also the diameter in inches of the indentation made by the hammer.

TABLE IV.

Bar.	Hardness Number (Shore). Mean of Twelve Determinations.	Diameter of Indentation. Mean of Twelve Determinations in Inches.
A	26	0.01535
B	26.3	0.01692
C	26	0.01830

From the hardness numbers obtained by the scleroscope it would be inferred that the three samples were of approximately the same hardness, but on measuring the indentation it is found that the hardest sample is A and the softest sample C, these determinations confirming the hardness taken by the drill method and quoted on page 59.

The following two diagrams (Figs. 5 and 6) show the Shore

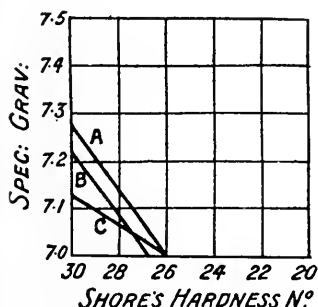


FIG. 5.

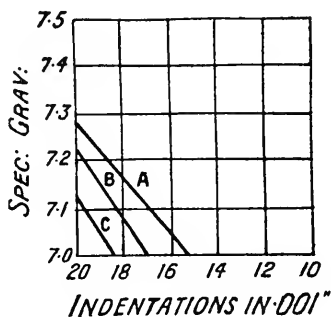


FIG. 6.

hardness number plotted against the density of the samples and the diameter of the indentation plotted against the density. It will be noticed that whilst the Shore number does not bear any direct ratio to the density, the diameter of the indentation does.

Sulphur Prints.

Sulphur prints were taken of the microsections of the bars A B and C, and it was observed that the sulphur was evenly distributed in the bars A and B, whilst in the bar C the

sulphur was found in segregated patches near the blowholes. This observation suggests that irons containing high sulphur should be cast at a high temperature, otherwise sulphur is eliminated from irons cast at a dull temperature, forming blowholes.

Gases Dissolved in the Samples.

The method adopted to obtain the gases dissolved in the samples has been described earlier. The following Table V. gives the amount of gas evolved from the samples in cubic centimetres per gramme of sample heated to a temperature of 950° C. for 12 hours.

TABLE V.

Sample.	Cubic Centimetres of Gas evolved per Gramme of Sample.
A	0·365
B	0·235
C	0·126

The chemical analysis of the gas evolved is shown in Table VI.

TABLE VI.—*Constituents per Cent.*

Sample.	CO ₂ .	CO.	H ₂ N ₂ , &c.
	Per Cent.	Per Cent.	Per Cent.
A	2·98	23·6	68·42
B	3·36	26·0	70·64
C	3·78	23·5	72·72

From the above it is noticed that the lower the temperature of casting the higher percentage of carbon-dioxide, hydrogen, and nitrogen, &c., are in the gas evolved, also that the carbon-monoxide is decreased.

2nd Series.

In this series seven sets of bars were cast at intervals of a few minutes.

As in series No. 1, the bars were cast from cupola metal of similar analysis and from the same ladle, but the difference in casting times were taken to a greater length.

The intervals of time between the casting of the seven bars is given in Table VII.

TABLE VII.

Bar.						Time Cast.	Interval in Minutes between the Casting of the "K" Bars after Bar AA.
						Hr. Min.	
AA	4 25	...
BB	4 26½	1½
CC	4 27½	2½
DD	4 29	4
EE	4 30	5
FF	4 31	6
GG	4 31½	6½

The analysis and specific gravity of each of the bars is shown in Table VIII.

TABLE VIII.

Chemical Constituents.	Bars.						
	AA	BB	CC	DD	EE	FF	GG
	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.
Total carbon	3.092	3.085	3.106	3.109	3.116	3.121	3.161
Graphitic carbon	2.321	2.285	2.346	2.397	2.404	2.526	2.616
Combined carbon	0.771	0.800	0.760	0.712	0.712	0.595	0.545
Silicon	1.318	1.346	1.318	1.262	1.272	1.339	1.237
Sulphur	0.126	0.145	0.128	0.124	0.124	0.121	0.120
Phosphorus	0.067	0.090	0.084	0.079	0.078	0.080	0.072
Manganese	0.412	0.523	0.432	0.400	0.400	0.390	0.370
Specific gravity of bars	7.273	7.298	7.250	7.227	7.213	7.141	7.016

From these analyses it will be noticed the total carbon in the whole series is practically constant. The combined carbon, however, is greatest in the BB sample, but gradually decreases with the reduction of the casting temperature. It is interesting to note the analysis of the BB sample; it contains the highest percentages of combined carbon, sulphur, and manganese, which are much higher in each case than in the AA sample which was cast 1½ minutes earlier. This shows that,

during the interval that elapsed, between the casting of AA and BB, the manganese sulphide in the molten mass had a tendency to rise towards the top, and therefore BB contained the top layer of metal containing the excess manganese sulphide. It will also be noticed, the sulphur and manganese decreased as the casting temperature was lowered.

The specific gravity is greatest in BB and the least in GG.

The appearance of the fractured "K" bars are shown in Fig. 7, Plate VIII. From the photographs it will be observed that AA, BB, and CC are perfectly sound, whilst DD, EE, FF, and GG exhibit contraction and blowholes. The unsoundness of structure is best seen in Fig. 8, Plate IX., which is a photograph of the polished sections of the fractures used for the measurement of the microconstituents. An attempt will be made later to give an explanation of the reason why sample EE contains the large contraction hole, whilst samples FF and GG only contain blowholes.

Hardness of Samples.

As in the previous series the hardness of the samples were determined by the Shore scleroscope. Table IX. gives the hardness number, also the diameter of the indentation in inches made by the hammer.

TABLE IX.

Sample.	Hardness No. (Shore). Mean of Twelve Determinations.	Diameter of Indentation. Mean of Twelve Determinations.
AA . .	26	0.01576
BB . .	25	0.01456
CC . .	26	0.01676
DD . .	26	0.01736
EE . .	25	0.01807
FF . .	26	0.01884
GG . .	25	0.01961

From the hardness numbers obtained by the scleroscope it would be inferred that the seven samples were of approximately the same hardness, but on observing the diameter of

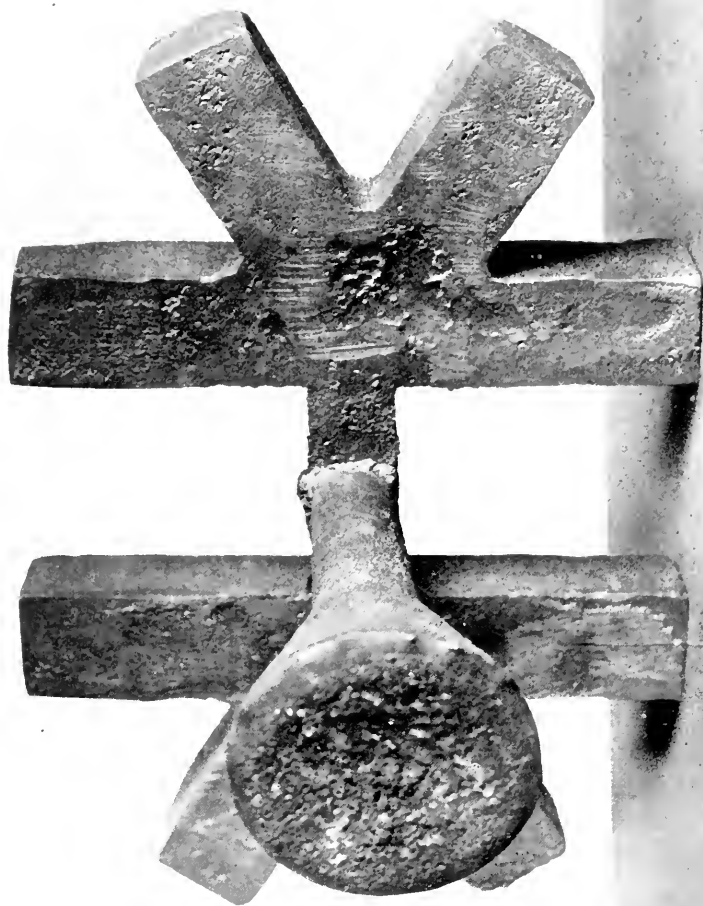


FIG. 3.





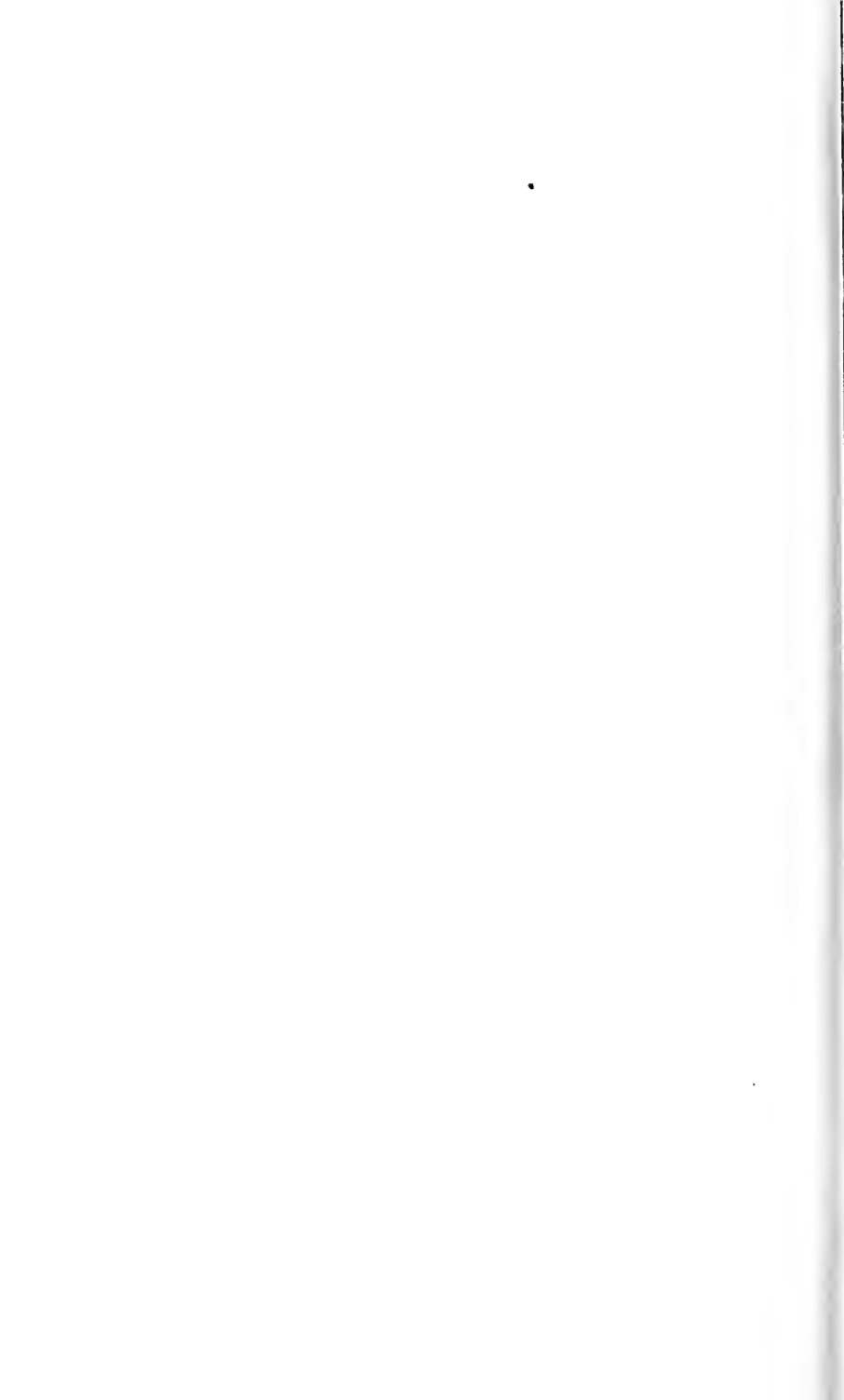
FIG. 4.

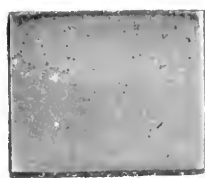




AA. BB. CC. DD. EE. FF. GG.

FIG. 7.

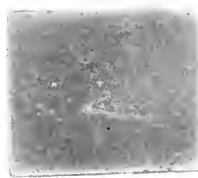




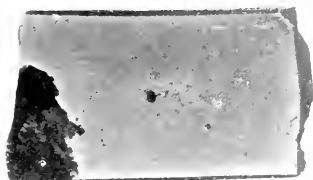
A A.



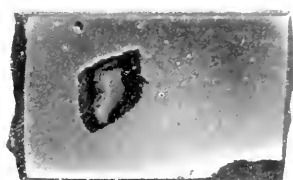
B B.



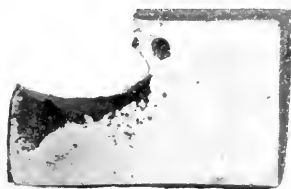
C C.



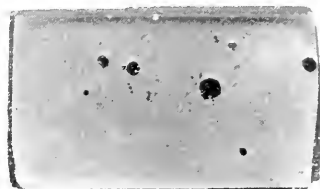
D D.



E E.



F F.



G G.

FIG. 8.



the indentation it will be noticed that different diameters are obtained which bear a relation to the specific gravity of the samples.

The two following diagrams (Figs. 9 and 10) show the Shore hardness number plotted against the specific gravity or density of the sample, and the diameter of the indentation plotted against the density.

Again it will be noticed that the Shore number does not bear any direct ratio to the density, although the diameter of the indentation does. This, the author thinks, proves the scleroscope to be of no value in the hardness testing of cast iron, as no reliance can be put on the hardness number recorded. It is of course conceded that the harder a sample the less will the hammer indent it.

Measurement of Constituents.

The samples AA, BB, and CC being of a sound nature, only one set of readings were taken, whilst samples DD, EE, and FF exhibit signs of liquid contraction, therefore two sets of readings were taken.

One set of readings was taken from the solid area of the samples, which is termed the "unaffected area," the other set being taken from the vicinity of the unsound portion, and termed the "affected area." The sample GG only was found to exhibit gas holes, therefore one set of readings only were taken. Table X. gives the mean dimensions of the constituents from twelve readings in each case. It will be observed that the sample BB has the highest density of the series, and contains the thinnest flakes of graphite and the largest areas of manganese sulphide. With reference to the samples DD, EE, and FF it will be noticed that in each case the graphite flakes are larger in the affected areas than in the unaffected areas, whilst the size of the manganese sulphide areas do not vary so much. On examining the Plate, Fig. 8, it will be seen liquid contraction is first observed in the sample DD, and is at its maximum in sample EE, decreasing in the sample FF, whilst the sample GG shows no sign, only blow-holes being present.

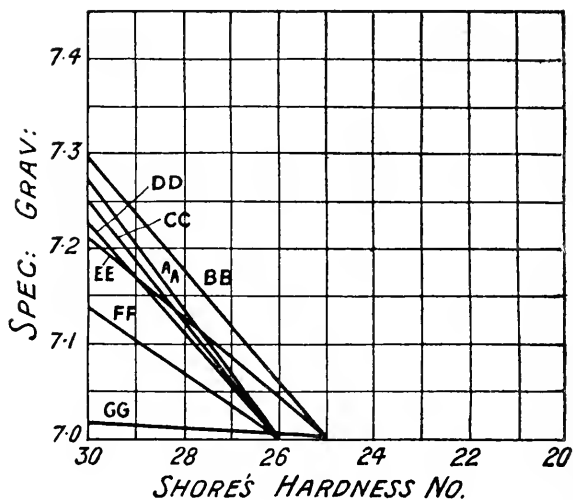


FIG. 9.

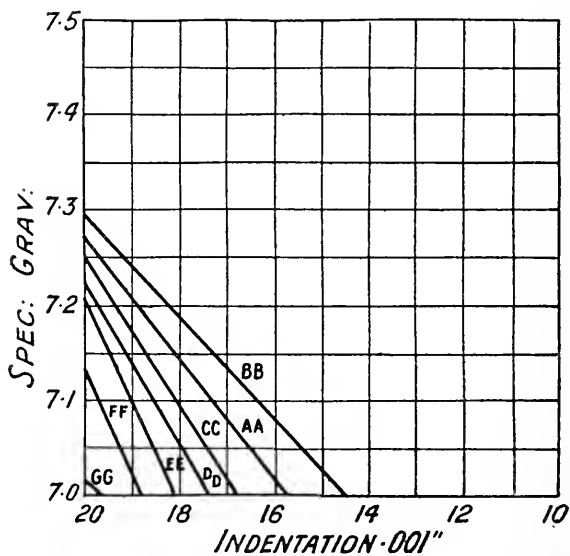


FIG. 10.

TABLE X.—Size of Microscopical Constituents.

Sample.	UNAFFECTED AREA.									
	Flakes of Graphite.			Plates of Graphite.			Manganese Sulphide Areas.			
	Length in Inches.	Breadth in Inches.	Length in Inches.	Breadth in Inches.	Area in Square Inches.	Length in Inches.	Breadth in Inches.	Area in Square Inches.	Length in Inches.	Area in Square Inches.
AA	0.002802	0.0000588	0.001147	0.000676	0.000000775	0.000284	0.000245	0.0000000696		
BB	0.002598	0.0000514	0.000922	0.000588	0.000000543	0.000441	0.000294	0.0000001297		
CC	0.003528	0.0000717	0.001077	0.000588	0.000000634	0.000250	0.000223	0.0000000575		
DD	0.002235	0.0001004	0.000676	0.000470	0.000000318	0.000259	0.000200	0.0000000518		
EE	0.002590	0.0000611	0.000685	0.000401	0.000000274	0.000259	0.000235	0.0000000609		
FF	0.001765	0.0000541	0.000629	0.000470	0.000000249	0.000187	0.000156	0.0000000292		
GG	0.002860	0.0000588	0.001019	0.000706	0.000000720	0.000153	0.000137	0.0000000209		
AFFECTED AREA.										
DD	0.00470	0.000117	0.00105	0.000588	0.000000618	0.000235	0.000215	0.0000000506		
EE	0.00435	0.000071	0.00160	0.001151	0.000001841	0.000225	0.000205	0.0000000664		
FF	0.00260	0.000058	0.00102	0.000617	0.000000629	0.000206	0.000171	0.0000000352		

On referring to Table X. again, it will be seen that the size of the graphite plates in the affected area of the sample possessing the greatest liquid contraction, *i.e.* EE, are by far the largest in the whole series. It was noticed in examining a large number of specimens containing these large contraction holes, that the plates of graphite in the immediate vicinity have always been found to be very large; it was therefore inferred that the hole had been formed, to a very large extent, by the action of the large plates of graphite on the constituents, to which it was in immediate contact, drawing them, as it were, to itself, and thus leaving a hole.

*Gases evolved on Heating the Specimens for Twelve Hours
at a Temperature of 950° C.*

Table XI. gives the amount of gas evolved from the specimens per gramme, also the analysis of the gas. Again it will be noticed, as in the series No. 1, that there is a gradual decrease in the volume of gas generated, and that the carbon-dioxide, nitrogen, hydrogen, &c., increase in percentage, and the carbon-monoxide decreases the colder the metal is cast.

TABLE XI.

Sample.	Cubic Centimetres evolved per Gramme.	Constituents per Cent.		
		CO ₂ .	CO.	H ₂ N ₂ , &c.
AA	0·392	1·80	25·26	72·94
BB	0·316	1·89	23·12	74·99
CC	0·253	2·15	24·68	73·27
DD	0·245	2·56	24·01	73·43
EE	0·225	3·18	23·96	72·86
FF	0·139	3·26	22·89	73·85
GG	0·096	3·36	22·51	74·13

It will have been noticed that in all the tests carried out with this second series, the bar BB has given better results than AA. This is undoubtedly due to the fact that AA was cast with metal too hot, being on the boil.

SUMMARY.

From the foregoing, and the results of many unrecorded but confirmatory tests, the author is of the opinion that:—

1. Liquid contraction is accompanied by a lowering of the specific gravity or density.
2. Casting temperature is one of the ruling factors in obtaining a dense and solid casting, and therefore also affects the liability to liquid contraction.
3. When castings have been cast with iron hotter than will conduce to liquid contraction, solid and dense castings result; when cast with iron too cold to conduce to liquid contraction, blowholes appear (see FF and GG, Fig. 8). Hence the series of conditions is:—Cast hot gives dense castings; medium heat, liquid contraction; cold iron, blowholes.
4. The more regular the sizes of the microscopical constituents the less will be the liquid contraction.
5. Conditions which conduce to a maximum solid contraction conduce to a maximum liquid contraction.
6. Gases appear to be held in more intimate contact in cast iron poured at a high temperature, and therefore tend to produce a more solid and dense casting.

This research has dealt with iron of the better variety as being the most liable to be affected, and also such as to have a tendency to cause the greatest worry and loss to the manufacturer in rejected castings. The author hopes, however, to continue the research with irons of the commoner variety.

THE ESTIMATION OF OXYGEN IN IRON AND STEEL.¹

By J. A. PICKARD, B.Sc. (LOND.), A.R.C.Sc.

IN order to arrive at a satisfactory method for estimating oxygen in iron and steel, it is first necessary to consider in what possible states of combination the oxygen may be present. All commercial varieties of steel contain—in addition to iron—carbon, silicon, manganese, sulphur, and phosphorus, and there is a theoretical possibility that oxygen may be combined with any or all of these, as well as of its being present in the free state.

The presence of oxides of carbon is rendered probable by their evolution, when iron and steel are heated to quite moderate temperatures *in vacuo*, as shown by the work of Parry, Graham, Boudouard, Goerens, Belloc, and others; and Goutal has also obtained similar results by dissolving steel in slightly acid copper potassium chloride solution. The presence of oxides of iron, manganese, and silicon is established by the presence of slag in the metal. This consists essentially of a mixture of ferrous and manganous silicates, and is probably formed as a result of the initial formation of the separate oxides, which afterwards combine to produce slag by reason of the chemical affinity of the basic and acid oxides. For several reasons it seems most likely that the iron oxide present is FeO . In the first place, it is reasonable to suppose that in the presence of such a very large excess of free iron the higher oxides, even if originally present as such, would be reduced to the lower. This is well known to be a fact in the case of copper, which always contains any oxygen present as cuprous oxide, even if originally added as cupric oxide. Reasoning from analogy is, however, always open to objection. In the second place, the experiments of Hilpert² show that

¹ Received March 11, 1913.

² *Berichte der Deutschen Chemischen Gesellschaft*, 1909, vol. xlii. 4893.

Fe_2O_3 on heating loses oxygen, forming successively FeO , Fe_2O_3 , 2FeO , Fe_2O_3 , &c., tending to a limit composition of FeO , which would be reached below the exceedingly high temperature of molten steel. In addition, slag always contains ferrous silicate, which indicates that ferrous oxide was originally present. Similar reasoning would lead to the conclusion that manganous oxide, MnO , would be present, but in this case fewer experimental data are available. The work of Matwieff¹ shows that ferrous oxide and manganous oxide are present together in steel in a state of very intimate admixture. The presence of oxidised phosphorus compounds is impossible, in view of their ready reducibility to phosphides. Sulphur dioxide is excluded by the fact that it has never been found in the gases evolved on heating *in vacuo*. The presence of free oxygen in steel or iron which has cooled from the molten condition is unthinkable, and the fact that oxygen is commonly stored under pressure in steel cylinders prohibits the likelihood of its being taken up through porosity by the cold metal. In addition to these constituents others are sometimes present. Of these, aluminium is the most common, in view of its utility in "killing" the molten metal. Vanadium, titanium, copper, and other metals are also occasionally present, and may be partially combined with oxygen in the finished steel.

It may be considered, therefore, that oxygen may be present in the forms of carbon monoxide, carbon dioxide, ferrous oxide, manganous oxide, and slag, and also in combination with any of the special constituents added. In the first four states of combination, its effect on the quality of the steel is not definitely proved, but is generally considered to be injurious. The oxygen in slag is known to be of small importance, as variations in the slag content do not greatly affect the mechanical properties. This is usually attributed to the insolubility of slag in steel, and there is a general agreement that those oxides which are insoluble cannot greatly affect the quality of the steel. Should this very reasonable theory prove correct, from the point of view of the steel-user and manufacturer, a method of estimating the soluble oxides,

¹ *Revue de Métallurgie*, 1910, 447.

while taking no account of the insoluble, would prove of greater practical value than a theoretically more satisfactory method which would estimate the whole.

The question of the solubility of oxides in steel is one of the most important at present occupying the attention of metallurgists, but a great deal of work remains to be done on the subject. The evidence at present available tends to show that ferrous oxide is soluble to a certain extent.¹ Manganous oxide is generally supposed to be insoluble, though Ledebur (in discussing Romanoff's paper²) has expressed an opinion that it is soluble to a slight extent. The same metallurgist has also stated that silica resulting from deoxidation remains suspended in small particles and is not dissolved: this, of course, is in cases where it does not combine with other oxides to produce slag. Aluminium oxide, and the oxides of the other special additions, appear to be at least as insoluble as manganous oxide.

The condition in which gases are present in steel is still a matter of controversy. Nitrogen and hydrogen, as is well known, can be obtained by boring steel under water or mercury, but the oxides of carbon are always absent from these gases, or occur in the merest traces. The latter are always found by igniting the steel *in vacuo*; and it is the author's experience, from some work on the subject that is not yet completed, that the smaller the particles of steel heated, the larger is the amount of gas—particularly CO and CO₂—obtained, notwithstanding that the more complete disintegration of the steel would result in a greater loss of mechanically held gas. These facts point to the conclusion that the oxides of carbon obtained, if originally present as gas, must have been held by a kind of "molecular porosity," as Deville has expressed it, which would amount to solution. There is also, of course, the alternative explanation, that the gases are produced by reduction of oxides by the carbon in the steel. The work of Goutal mentioned above controverts the belief that they are entirely due to this cause, but it would be very interesting to have more evidence on this point.

¹ Law, *Journal of the Iron and Steel Institute*, 1907, No. 11. p. 99.

² *Ibid.*, 1899, No. 1. p. 450.

The method worked out by the author provides for the estimation of ferrous oxide, manganous oxide, and oxides of carbon, the only oxides occurring in iron and steel for which there is any evidence of solubility.

PREVIOUS METHODS FOR THE ESTIMATION OF OXYGEN.

Pourcel's Method.—This consisted in heating the steel, in the form of drillings, in a stream of chlorine, until all the iron was volatilised and driven off as FeCl_3 . The residue was weighed after removal of carbon. Such obvious objections as the doubtful composition of the residue and the assumption that the oxides are unaffected by the treatment, as well as the difficulty of manipulation, make this method unreliable. One of the most important sources of error is the reduction of the oxides by the carbon, which is not volatilised by the chlorine, resulting in loss of oxygen as carbon monoxide and of the corresponding iron as ferric chloride. The importance of this loss may be judged from the fact that to heat a metallic oxide with carbon in a current of chlorine is a stock method for preparing metallic chlorides. The method has been severely criticised by Ledebur,¹ who states that this method showed "about half the oxygen that was present in the iron," which presumably means about half that shown by Ledebur's own method.

Hydrogen Reduction Methods.—Bender² appears to have been the first to describe a method for estimating oxygen by simple heating in hydrogen. The same method with modifications was used by Ledebur,³ and Dr. Stead has informed the author that a similar method was employed in his laboratories many years ago. This method has so far proved the most satisfactory, and merits a fairly full description. As Ledebur's own original details were not accessible, the following account is taken from Max Orthey's *Eisenhüttenchemie*,⁴ in which the method is reprinted.

"About 15 grammes of steel are placed in an ignited

¹ *Journal of the Iron and Steel Institute*, 1895, No. II. p. 580.

² *Dingler's Polytechnisches Journal*, 1872, vol. ccv. p. 531.

³ *Leitfaden für Eisenhütten Laborationen*, 6th Edition, p. 122.

⁴ 1907, p. 199.

porcelain boat and introduced into the combustion tube. Before attaching the weighed absorption tube, the tube is swept out with pure dry nitrogen until all air is removed. It is then heated for a considerable time to redness in the same stream, to get rid of all traces of air remaining in the drillings and to decompose organic compounds. The tube is then allowed to cool in the stream of nitrogen, and the nitrogen replaced by hydrogen. The weighed absorption tube of P_2O_5 is then attached, followed by a guard tube of H_2SO_4 , and the tube is then heated to bright redness for 30 to 40 minutes. The tube is then allowed to cool slowly, hydrogen replaced when cold by pure dry air, and the absorption tube weighed. The increase in weight, multiplied by 8/9, will equal the oxygen. As a check, the turnings and boat may be weighed; but as H_2S is lost, the loss in weight will always be greater than the gain of the absorption tube."

There are several errors involved in this mode of procedure. During the first heating in nitrogen a large proportion, if not all, of the carbon dioxide and carbon monoxide contained in the steel will be evolved and lost, and there is also the possibility, if the heating is at all prolonged, that some of the oxides present may be reduced by the carbon in the steel, causing a further loss. The reduction of oxygen-containing constituents of steel by carbon contained in the metal is known to take place pretty readily at 900–950° C. A familiar example is the production of the blisters in blister steel, which are caused by the evolution of carbon monoxide resulting from the reduction. A loss from this cause is particularly liable to occur with high carbon steels. Also, during this stage any hydrogen which is present in the steel may combine with some of the oxides, causing a further loss; while to heat steel to redness is hardly to be recommended as a method for removing traces of mechanically entangled oxygen. Finally, it is the author's experience that glazed porcelain ware reacts fairly readily with turnings of steel which may be heated in it. During the actual reduction in hydrogen there is a probability of further loss in consequence of the fact that carbon monoxide will again be produced by the reduction of some of the oxides by carbon present.

Carbon monoxide is only reduced with difficulty by hydrogen. Gautier ¹ finds that when dry hydrogen and carbon monoxide are passed through strongly-heated porcelain tubes, water and carbon dioxide are formed. When passed in the proportion of 3 volumes of hydrogen to 1 volume of carbon monoxide at 1220° C., which was the temperature producing the maximum yield of water, only 2.75 per cent. by volume of the carbon monoxide was reduced. Using the same mixture and passing rapidly at 1300°, 8 per cent. of the carbon monoxide was reduced to methane, but this figure could not be exceeded. It is evident from these figures that at the lower temperature of the estimation of oxygen, even in the presence of the much larger excess of hydrogen, the major part of any carbon monoxide formed would escape transformation into water. In addition to these objections, the method has the disadvantage that a complete working day is needed for each estimation. Romanoff ² and Cushman ³ shorten the procedure by omitting the preliminary heating in nitrogen, but Law ⁴ finds that this leads to irregularity in the results.

Method of Troilius.—This method ⁵ depends upon treating the steel with a neutral solution of ferric chloride. When the reaction is complete, the residue, consisting of sulphide, phosphide, and oxide of iron with traces of silica, is filtered off, and washed with water and dilute hydrochloric acid, ignited and weighed, and the silica allowed for. Although this method possesses the advantage of weighing the oxygen in the form of ferric oxide, which is considerably heavier than the water which would be formed from the same amount of oxygen, the author has not succeeded in proving that the oxide is not attacked and, partly at least, removed by the action of the reagents. Ferrous oxide is a remarkably active compound. When prepared by heating iron in carbon dioxide to about 1000°, it decomposes water at ordinary temperatures and is pyrophoric; ⁶ it also readily dissolves in acids to form ferrous salts.

¹ *Comptes Rendus*, 1910, pp. 150, 1564.

² *Stahl und Eisen*, vol. xix. pp. 265-269.

³ *Journ. Ind. Eng. Chem.*, vol. iii. p. 372.

⁴ *Journal of the Iron and Steel Institute*, 1907, No. II. p. 99.

⁵ *Ibid.*, 1885, No. I. p. 330.

⁶ Moissan, *Ann. Chim. Phys.* [5], vol. xxi. p. 199.

Other Wet Methods.—Several other wet methods have from time to time been put forward. Solution in bromine water or iodine water, or in copper potassium chloride solution containing bromine, followed by ignition and weighing of the residue, have all been proposed, but are all open to the same objection of doubtful composition of the residue and uncertainty that the ferrous oxide is unaffected by the reagents. Tonnele and Carnahan, however,¹ who have examined a large number of methods, consider that the last named gives the best results.

Tucker's Method.—This consists in melting about a kilogram of steel, the weight of which is accurately known, in a crucible lined with charcoal, taking precautions to prevent the oxidation of the surface.² After a sufficient period of fusion, the steel is allowed to cool again, and is reweighed. The percentage of carbon is estimated before and after melting. The loss in weight, after allowing for the carbon taken up, represents the oxygen removed. The disadvantages of this method as a routine estimation of oxygen are considerable, and its accuracy has been severely criticised by Romanoff (*loc. cit.*). In addition, the exceptionally high results obtained by its use render it liable to suspicion.

Müller's Method.—The amount of oxygen in blown metal is determined by estimating the loss of carbon, manganese, and silicon in the deoxidation of the metal. The results may be interpreted by two hypotheses as to the course of the reactions followed, and the resulting figures vary widely according as one or the other is adopted. In one case where the first hypothesis gave 0.089 per cent., the second gave 0.394 per cent. of oxygen. This method was not intended as a laboratory estimation, and indeed could not be made general, as the deoxidation of steel by manganese is now generally allowed to be a balanced reaction depending on the amount of manganese added.³

¹ *Proceedings of the Engineers' Society of Western Pennsylvania*, vol. xii. p. 295.

² *Journal of the Iron and Steel Institute*, 1881, No. I. p. 205.

³ Müller, *Zeit. Vereins. Deutsch. Ingenieure*, xxii. p. 385; Howe, *Metallurgy of Steel*, p. 93.

OUTLINE OF THE PRESENT INVESTIGATION.

When the present investigation was undertaken, the author had in view two possible methods. In the first it was proposed to melt steel *in vacuo* in a graphite crucible for a length of time sufficient to reduce all the oxides, with formation of the equivalent amount of carbon monoxide. It was intended subsequently to extract the carbon monoxide and estimate it by passing it over warm iodine pentoxide, which would result in the production of an equivalent amount of iodine which could be readily titrated. The second was a modification of the hydrogen reduction method which would eliminate the errors involved in the former modes of operation. Work on the first method was still in the preliminary experimental stage when Walker and Patrick, of the Massachusetts Institute of Technology, published their method for the estimation of oxygen in steel.¹ This method only differs in a few experimental details from the author's projected method, and further experiments were therefore discontinued, and the second method taken up instead.

DETAILS OF THE METHOD USED.

Hydrogen free from oxygen is generated in a Kipp's apparatus (not shown in diagram) by the action on zinc of dilute sulphuric acid (1 : 6, free from nitric acid and oxides of nitrogen, and containing about 5 grammes of crystallised ferrous sulphate per 100 cubic centimetres). The gas is purified by bubbling through strong sulphuric acid in the washing bottle A, from which it passes to B, containing solid caustic potash, followed by calcium chloride. It is then thoroughly dried by passing through the phosphorus pentoxide in C. All rubber connections are rendered perfectly gas-tight by a covering of "Chatterton's compound" or sealing-wax. E is a vitreous silica tube closed at one end, 18 inches long by 1 inch in diameter, the walls being 3 mm. thick; it is connected by a gas-tight rubber joint with the glass extension F. This is of the same

¹ *Eighth International Congress of Applied Chemistry*, 1912, vol. ii. p. 139.

diameter, carries a side tube L, and is closed by a glass cap H, which fits over the end. The side tube L carries two stop-cocks, D and M, on a T-piece. D is connected with the hydrogen supply, and M with a Fleuss pump. The silica tube is held horizontally in a strong clamp in such a manner that an electric furnace can be placed round it for nine or ten inches of its length, and can be removed at will without disturbing any part of the apparatus. The steel to be analysed is placed in a nickel boat (previously ignited in hydrogen) and pushed up near the closed end of the tube. The water resulting from the reduction is absorbed by a small glass boat

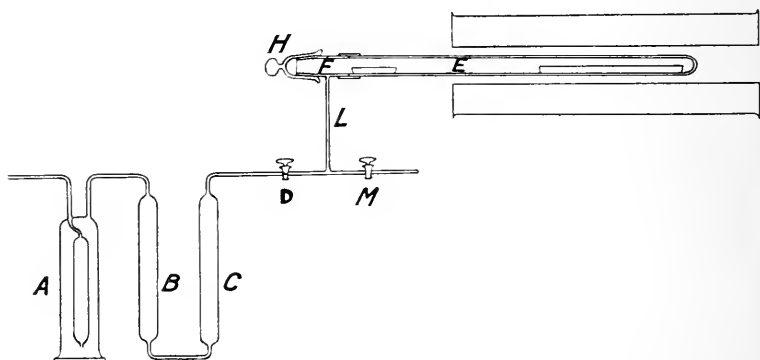


FIG. 1.

containing phosphoric oxide inserted into the open end of the tube near the glass cap.

To carry out a determination, the silica tube is first completely freed from moisture by carrying out a blank heat with an unweighed boat of phosphoric oxide in the cold end. When cold the nickel boat is taken out and weighed. About 20 grammes of steel are then placed in the boat and the whole is reweighed. The steel should be in the form of drillings or turnings, and preferably free from moisture—this is readily ensured by keeping the drillings in a desiccator. A small glass boat containing phosphoric oxide is weighed in a special weighing bottle provided with small feet to prevent rolling on the balance pan. The drillings are now placed in the

silica tube near the closed end and a small unweighed boat containing phosphoric oxide placed near the open end. The cap is replaced and the tube evacuated and allowed to remain so for a minute or two, during which time any moisture and air condensed on the drillings is completely removed. Hydrogen is now admitted up to atmospheric pressure, and the weighed boat of phosphoric oxide substituted for the other. While the change is being made, the hydrogen completely escapes from the tube and is replaced by the same volume of laboratory air. A small correction is made for the amount of water vapour introduced in the air, but this never amounts to more than 1 or 2 milligrammes. The tube is now exhausted, filled with hydrogen, and re-exhausted. In order to remove the last traces of air, this washing out with hydrogen may be performed twice. Finally, hydrogen is admitted up to half or two-thirds atmospheric pressure, which is readily effected by arranging that the volume of the purifying apparatus between tap D and the tap on the Kipp's apparatus is from one to two times that of the silica tube and attachment. Then, the purifying apparatus being filled with hydrogen at slightly more than atmospheric pressure, by shutting off the Kipp's apparatus and making connection with the vacuum tube, the latter is filled with hydrogen at the right pressure. The object of this diminished pressure is to prevent the cap being blown off by the increased pressure due to the expansion of the gas on heating. The electric furnace—previously heated to 1000° —is now pushed over the end of the tube until 9 or 10 inches are surrounded, and allowed to remain there for three-quarters of an hour, or less if the drillings are fine. At the end of this time the furnace is removed and the tube allowed to cool. The cooling is greatly assisted by an air-blast, but this is not essential. When cold it is filled with hydrogen up to atmospheric pressure, tap D closed, the cap removed, and the P_2O_5 boat quickly replaced in the weighing bottle and weighed. The loss in weight of the steel is always slightly greater than the gain of the P_2O_5 boat by reason of the loss of sulphur and carbon. A complete estimation may be carried out in less than an hour, and during most of the time the operator is free to do other work.

Determination of Correction due to Moisture in the Air introduced.—This may be done either by simply carrying out a blank heat, or by a separate estimation of humidity in the air—the second way being quicker—the results by both methods agreeing closely. The estimation of humidity is most conveniently carried out as follows: the bulb A (Fig. 2), fitted with a three-way capillary tap B, contains 100 cubic centimetres of gas between the tap and the mark C, and may be filled

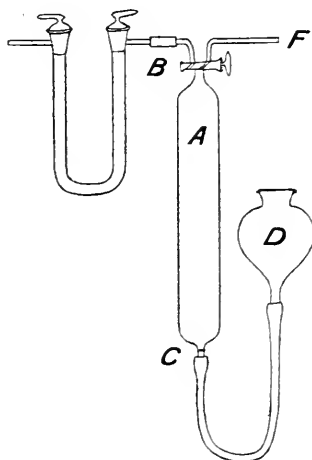


FIG. 2.

with mercury by raising the reservoir D. A weighed absorption tube of P_2O_5 is attached as shown. One hundred cubic centimetres of laboratory air are drawn in through F by lowering the reservoir and then gently expelling through the absorption tube. When 500 cubic centimetres have been driven through, the tube is removed and re-weighed—the increase being the weight of water in 500 cubic centimetres of air. The correction to be applied is found from the relation of the volume of the tube and the weighing bottle to 500 cubic centimetres. Thus—

$$\frac{\text{Volume of weighing bottle} + \text{volume of tube}}{500} \times H_2O \text{ found} \\ = \text{correction.}$$

This is never greater than 1 or 2 milligrammes.

EXPERIMENTAL RESULTS.

Determination of Blanks.

The following figures show the agreement of blanks as determined by both methods, and the necessity for washing out the tube twice with hydrogen:—

No. of Washings.	H ₂ O obtained.
1	0.0020 gramme.
2	0.0012 „
3	0.0012 „

Weight of water in 500 cubic centimetres air=0.0034 gramme.

Blank calculated for tube (174 cubic centimetres)=0.0011 gramme.

Reduction of Carbon Dioxide and Carbon Monoxide.

The following experiments were performed to determine the conditions under which the oxides of carbon are reduced by hydrogen, with formation of the equivalent amount of water.

A known amount of carbon dioxide was obtained by weighing pure calcium carbonate into the nickel boat and introducing it into the silica tube. The reduction is difficult, and only takes place completely when a large excess of hydrogen (100 to 1 by volume) is present. A temperature of at least 1000° C. is necessary, and the heating must occupy at least half an hour. Temperatures as high as 1200° C. have been used but not found to offer any marked advantages; but longer heating compensates to some extent for lower temperatures. It is to be noticed that when higher amounts of carbon dioxide than about 2 cubic centimetres are taken, complete reduction to water does not result. As the tube contained about 110 cubic centimetres of hydrogen at atmospheric pressure, this corresponds to nearly 2 per cent. by volume. The maximum amount of CO₂ and CO to be expected from 10 grammes of any ordinary steel will not exceed 2 cubic centimetres, so that under the conditions of experiment no loss of oxygen through incomplete reduction of carbon monoxide is to be feared. Steel—previously ignited in hydrogen—was present in some experiments, but had no beneficial effects on the reduction. The following table shows some typical experiments:—

Experi- ment.	Details.	Weight CaCO ₃ .	Repre- senting CO ₂ .	O found.	O present as CO ₂ .	Percentage of oxygen as CO ₂ reduced to H ₂ O.
1	30 minutes' heating at 1200° C.	Grm. 0·0729	c.c. 16·3	Grm. 0·0149	Grm. 0·0233	64·0
2	Same. Steel present	0·2430	54·3	0·0487	0·0779	62·6
3	15 minutes at 1100°	0·0490	11·0	0·0069	0·0156	44·2
4	30 minutes at 1200°	0·0866	19·4	0·0167	0·0276	60·5
5	60 minutes at 1050–1100°	0·0641	14·4	0·0128	0·0205	62·5
6	30 minutes at 950–1000° Pressure of hydrogen, atmospheric	0·0081	1·8	0·0026	0·0026	99·7

In experiments 1–5, the pressure of hydrogen was about 500 millimetres Hg when the tube had attained its full temperature. In experiment 6, the pressure was about 780 millimetres.

Reduction of Ferrie Oxide by Hydrogen.

Carefully dried rouge was used in this experiment. Reduction was complete in less than a quarter of an hour at 900°.

(i.) 0·0941 gramme rouge gave 0·0319 gramme H₂O, representing 0·0284 gramme oxygen. Oxygen present, 0·0282 gramme.

(ii.) 0·0800 gramme rouge gave 0·0276 gramme H₂O, representing 0·0245 gramme oxygen. Oxygen present, 0·0241 gramme.

Reduction of Manganese Oxide by Hydrogen.

For these experiments a specimen of peroxide of manganese containing 89·8 per cent. MnO₂ was used. Simple ignition in hydrogen only resulted in a reduction to manganous oxide, nor was any further reduction produced by igniting in contact with Swedish iron drillings which had themselves been previously ignited in hydrogen.

Weight of Sample.	Conditions of Experiment.	H ₂ O found.	H ₂ O resulting from reduction of MnO ₂ to MnO.
Gramme. 0·3081	Simple ignition in H	Gramme. 0·0558	Gramme. 0·0572
0·2777	Same	0·0526	0·0515
0·1982	Ignition in presence of reduced Swedish iron	0·0371	0·0368

Notwithstanding these results, the work of Matweieff (*loc. cit.*) shows that MnO mixed with FeO, as it occurs in steel, is readily reduced to manganese by hydrogen at 300° C.

Reduction of Slag by Hydrogen.

Two samples of ladle slag only were examined, and neither of these showed any reduction either when simply ignited in hydrogen, or when ignited in hydrogen in contact with reduced Swedish iron.

Thickness of Sample.

As in the determination of carbon by direct combustion in oxygen, the thickness of the drillings or turnings used has a very important effect on the percentage of the constituent found. As a practical means of deciding the limit of thickness permissible, a uniform billet of steel was turned down in the lathe, taking cuts of varying depth, afterwards measuring the thickness of the shavings with a micrometer gauge. This method does not admit of a very close approximation to the upper limit, but the results show that this is above 0.2 millimetres. The most satisfactory form of sample for analysis is turnings of 0.2 millimetre ($\frac{1}{125}$ inch) thickness, or less. In practice, if drillings are clean it is sufficient to sift out the finest part through a 30-mesh sieve and work on these, but in every case the drillings or turnings must be taken with every precaution to ensure absence of scale, rust, and oil; and, of course, the drillings or turnings must not be taken off hot.

Sample 1.—Basic open-hearth billet containing 0.55 per cent. C.

Experiment.	State of Sample.	Oxygen found.
		Per Cent.
1	Turnings about 0.1 millimetre thick	0.019
2	Same	0.020
3	Turnings about 0.15 millimetre thick	0.019
4	Same	0.017
5	Turnings about 0.7 millimetre thick	0.013
6	Same	0.008
7	Fine sifted portion from drillings	0.020

Sample 2.—Another basic open-hearth billet, containing carbon 0.17 per cent., manganese 0.38 per cent., silicon 0.025

per cent., sulphur 0.025 per cent., and phosphorus 0.022 per cent., gave the following figures:—

Experiment.	State of Sample.	Oxygen found.
		Per Cent.
1	Average sample from drillings	0.014
2	Fine sifted portion	0.025
3	Same	0.026
4	Thickest part of drillings	0.002
5	Finest drillings obtainable	0.027

Using a temperature of 1200° instead of 1000° did not alter the percentage of oxygen found in the thick samples. The results with turnings of varying thickness show that the higher figures obtained when using the fine sifted, in preference to the coarse, drillings, are not due to variation in composition of the sample.

Absorption of Oxygen from the Air by Drillings.

Oxygen from the air is only taken up to a small extent by drillings, if at all. Drillings weighed immediately after cooling from the ignition in hydrogen when allowed to remain in the balance case overnight never increased in weight by more than 0.0015 gramme on 20 grammes of steel. This increase was completely removed by another evacuation in the silica tube in the presence of P_2O_5 —that is, by the first operation performed in the method of estimation of oxygen.

Oxygen Content of various Steels.

The author hopes to be able to make a much fuller examination of the oxygen content of various steels and irons than has been possible so far, the object of this research having been to work out a satisfactory method of estimation. The following results were obtained while testing the applicability of the method. One sample only of each steel was analysed, and while the results are probably representative, it is not suggested that all steels made by the same process and with the same carbon content would give the same figures. The first five steels are by the same maker, and similar except for carbon content. Number 6 is by the same process but a different maker—the same sample as No. 2 above.

The cast iron, No. 8, is from a casting of grey iron. The Swedish bar iron, No. 7, contained carbon 0.013 per cent., manganese 0.025 per cent., silicon 0.01 per cent., and was of the quality generally used for standards as 99.9 per cent. iron. The amount of oxygen found in the last is startling, being more than four times as great as the total amount of other impurities.

Sample.	Carbon	Process of Manufacture.	Oxygen present as Oxides.
	Per Cent.		Per Cent.
1	0.10	Basic open-hearth	0.038, 0.036
2	0.20	Same	0.031, 0.028
3	0.30	Same	0.030, 0.027
4	0.44	Same	0.029, 0.025
5	0.55	Same	0.020, 0.020
6	0.75	Same	0.025, 0.026, 0.027
7	0.013	Swedish bar iron	0.358, 0.352
8	...	Soft grey cast iron cast in foundry	0.009, 0.006

SUMMARY.

A brief review is given of our knowledge of the compounds containing oxygen which occur in iron and steel. Methods used by other workers for the estimation of oxygen are critically examined and causes for their failure adduced. These causes are briefly—(1) For wet methods: the uncertainty of the action of the reagents used on the oxides present; (2) for dry methods: neglect of the reducing effect of carbon, disregard of oxygen contained as carbon dioxide and monoxide, and imperfect elimination of air and moisture. A new method is described depending essentially on the ignition of the steel in a confined volume of hydrogen containing also a weighed amount of phosphorus pentoxide. The precautions to be observed in performing the estimation are given, together with an examination of the effect of various factors—namely, temperature, length of heating, thickness of sample, and air absorption.

In conclusion, the author wishes to thank Dr. F. Rogers for the interest he has taken in the work, and also for providing him with some of the materials; and also Mr. A. W. Gregory for kindly supplying him with a number of samples of steel.

INFLUENCE OF INTERCRYSTALLINE COHESION UPON THE MECHANICAL PROPERTIES OF METALS.¹

By J. C. W. HUMFREY, B.A., M.Sc., M.ENG.

IN the course of the last few years several writers have advanced a theory that the irregularly outlined crystals of which metals are built up are joined to one another by thin layers of metal in an amorphous state. Experimental evidence in support of this theory has been published by Bengough² and by Rosenhain and Ewen.³

The latter authors, in discussing the existence of an amorphous intercrystalline cement, advance a theory as to the causes which bring about its formation. While fully accepting the existence of such a cement (and indeed some recent researches upon the mechanical properties of mild steel at high temperatures carried out by Rosenhain and the present author⁴ provide experimental data which can only be explained on some such hypothesis), the views presented below differ in certain details regarding its formation and structure from those advanced by Rosenhain and Ewen.

The present report deals with the effects which the amorphous intercrystalline cement has upon the mechanical properties of metals at ordinary temperatures, and may be conveniently divided into three sections:—

1. A theory as to the formation and structure of intercrystalline junctions.
2. Some microscopical observations upon plastic strain in metals, which offer certain evidence in support of this theory, and suggest

¹ Received March 18, 1913.

² Bengough, "A Study of the Properties of Metals at High Temperatures," *Journal of the Institute of Metals*, 1911, No. II. vol. vii. p. 176.

³ Rosenhain and Ewen, "Intercrystalline Cohesion in Metals," *Journal of the Institute of Metals*, 1912, No. II. vol. viii.

⁴ Rosenhain and Humfrey, "The Tenacity, Deformation, and Fracture of Soft Steel at High Temperatures," *Journal of the Iron and Steel Institute*, 1913, No. I.

3. An explanation of some of the mechanical properties of metals, and particularly those of the alterations in the elastic limits after overstrain.

(1) THE FORMATION AND STRUCTURE OF INTERCRYSTALLINE JUNCTIONS.

The essential condition of the crystalline state in matter is now generally recognised to consist in the regular and homogeneous arrangement of its ultimate units. It is further known that this arrangement is identical for any particular kind of matter under given conditions. Thus it has been shown that the centres of gravity of the molecules of each individual crystal are arranged according to one of fourteen geometrical designs (corresponding to the known crystal "systems") which are characterised by the property that every point is surrounded by a precisely similar arrangement of other points. Further than this, the work of Tutton has shown that the atoms composing the molecules are similarly situated in each molecule within the crystal. Pure metals crystallise, with few exceptions, in one of two systems possessing high symmetry, viz. the cubic and hexagonal, and since there are grounds for assuming that they are often mon-atomic, their structure may in such cases be considered as an assemblage of these atoms according to one of the two systems of closest packing.¹

In a metal solidifying from the molten state, the growth of crystals commences from nuclei whose number is apparently dependent both upon the purity of the metal and the rate at which it is cooled, and whose distribution within the crystal mass is entirely random. From these nuclei the individual crystals grow outwards (usually by means of dendrites) until further growth is prevented by interference with the sides of the mould or with their neighbours. In the molten state the molecules are without any definite order or marshalling, but it has been suggested by Barlow² that surrounding each

¹ Barlow and Pope, "The Relation between the Crystalline and Chemical Constitution of Inorganic Substances," *Transactions of the Chemical Society*, 1907, Part II. p. 1159.

² Barlow, "A Mechanical Cause of the Homogeneity of Crystals," *Scientific Proceedings of the Royal Dublin Society*, vol. viii. p. 264.

growing crystal there exists a layer of molecules which, although not held together in the rigid solid condition, are yet tending to a regular formation such as exists within the crystal itself. Now when two crystals are gradually growing towards one another, each will endeavour to marshal the still liquid molecules surrounding it to its own orientation; and it would seem, provided equilibrium could be established before the temperature fell below the point at which crystallisation becomes difficult, that there would be formed between the two crystals a layer of matter in which the molecules are so arranged as to pass by gradual small displacements from the orientation of one crystal to that of the other. If the cooling of the metal be very rapid, it is possible that before this state of complete equilibrium is achieved, the layer of liquid between the two crystals would become so viscous as to remain in a purely amorphous or undercooled state; but even in such a case the amorphous layer would be connected to each crystal by molecules whose distribution gradually approached that of the orientation of the crystal nearest them. Subsequent annealing of such a specimen would tend to reduce the layer of purely undercooled liquid and bring about the gradual and continuous change of orientation representing complete equilibrium.

In a previous report¹ the author has suggested that iron undergoes at least one complete recrystallisation in the solid state, viz., on cooling through Ar_3 , and this fact is confirmed by the subsequent work on high temperature deformation referred to above. In passing from the γ to the β condition the crystals change both their internal structure, and with it their individuality and outlines. The temperature of this recrystallisation is, however, still far above that at which crystal growth becomes impossible (since Goerens² has shown this point to be in the neighbourhood of $500^\circ C.$), and an adjustment of the molecules between contiguous β crystals can still proceed for some considerable range of temperature. It seems probable that in passing from the γ

¹ Humfrey, "The Intercrystalline Fracture of Iron and Steel," *Iron and Steel Institute: Carnegie Scholarship Memoirs*, 1912.

² Goerens, "On the Influence of Cold Working and Annealing on the Properties of Iron and Steel," *Iron and Steel Institute: Carnegie Scholarship Memoirs*, 1911.

to the β condition the structure must become at least momentarily amorphous, and thus the adjustment would still, as in the case of solidification from the liquid, be between crystal, undercooled liquid and crystal. The fact that the liquid would be very viscous might make the establishment of complete equilibrium more difficult, and cause a finite layer of amorphous matter always to remain between the crystals unless the cooling was particularly slow. This may account for some of the differences in the form and configuration of the slip bands found when iron is plastically strained, and those upon other metals which have undergone no such recrystallisation in the solid. Metals whose structure has been formed by recrystallisation in the solid, by annealing after severe strain, show similar slip bands to those found in iron.¹

Whether the structure be formed by direct crystallisation from the molten state or by recrystallisation in the solid, it thus appears to the author that while between the crystals of a metal there must exist a layer of material which does not possess the regularity of the crystalline structure and may therefore be considered as amorphous, yet the change from crystalline to amorphous is not abrupt, but consists of a gradual and continual displacement of the molecules from their regular positions in the crystalline orientation. If equilibrium between two adjacent crystals be complete, the matter between them consists of units so arranged as to pass by small displacements directly from the orientation of one crystal to that of the next; but if, from rapid cooling or other causes, true equilibrium has not been established, the crystals will be separated by a layer of metal in a completely amorphous condition.

In the condition of complete equilibrium, the thickness, in any particular plane of the region between the two crystals in which displacements from either true orientation occur, will depend upon the relative difference of orientations in this plane—the greater the difference of orientation, the more adjustment required, and hence the greater thickness necessary to complete it. In the plane in which the units are similarly orientated in either of two adjacent crystals, no displacements

¹ Rosenhain, "The Plastic Yielding of Iron and Steel," *Journal of the Iron and Steel Institute*, 1904, No. I. pp. 335-371.

of either orientation will be required in passing from one to the other, and if this plane be also one of gliding or cleavage, it might be expected that these properties would be continuous between the two crystals. Such cases may be found in the microscopic study of strain phenomena, and will be referred to later. In cases where the gliding and cleavage planes are not similarly oriented, since these properties are due to the regular and homogeneous structure of a crystal and the displacements in the orientation occurring between two crystals disturb this regularity, they will not readily be continued from one to the other.

The above considerations apply particularly when the meeting crystals are of similar composition, such as would be the case in a pure metal or one consisting of crystals of a single uniform solid solution. We can, however, imagine a similar boundary between crystals of different composition such as occur in alloys in which the constituents do not form a complete series of solid solutions. The gradual change of orientation would then be accompanied by a similar gradual change in composition; as the orientation of each crystal becomes disturbed, so its resemblance to the liquid, and hence its solubility for the other constituent, becomes more perfect. The actual thickness of the disturbance between two adjacent crystals of different structure would probably be greater than that between two of similar structure, and this supposition may account for the superior resistance which certain eutectics offer to plastic strain.

(2) OBSERVATIONS UPON PLASTIC STRAIN IN METAL.

The mode of deformation of the crystals in a specimen of metal subjected to plastic strain has now conclusively been shown to take place by means of slipping upon certain gliding planes, such slipping causing the well-known appearance of slip bands upon a previously polished surface. The following observations have been made with the idea of studying the behaviour of these slip bands at or near the intercrystalline boundaries, and thereby determining the influence which

these boundaries exert upon plastic deformation. The material first selected for the experiments was a pure Swedish sheet iron.

If a polished and lightly etched strip of this material be strained in a small machine (such as that originally used by Ewing and Rosenhain¹) so that the surface can be observed under the microscope while the straining process is being carried out, the following phenomena may be witnessed:—The slip bands which first appear when the elastic limit has been passed are confined almost entirely to the central parts of the crystals, and only spread gradually towards the boundaries as the straining becomes severe. The surface of each crystal tends to become curved as the straining proceeds, and the "ruffling" so produced renders a satisfactory focussing and illumination of the surface increasingly difficult. Figs. 1 and 2, Plate X., are micrographs taken from the same area of a specimen of iron after slight and after more severe straining. In Fig. 1 the commencement of slip in the centre of the large crystal in the middle of the micrograph is visible, while in Fig. 2 these slips have extended and others appeared, and considerable ruffling of the surface has taken place. It will be noticed that where a series of slip bands are parallel to a portion of a boundary they are broad, and the system continues close to it without the slip bands losing their direction, but that where they approach a boundary at an angle they become narrower, and tend to bend round so as to approach it at a smaller angle. Fig. 3, taken from another specimen, also illustrates these observations. In Fig. 4, which shows another part of the specimen illustrated in Fig. 3 at a higher magnification, the behaviour of the slip bands quite close to the crystal boundaries can be observed in greater detail, and the narrowing and bending round of the bands is more clearly visible. The almost complete absence of slip bands near some of the boundaries, as compared with the central parts of the crystals, is very marked in this photograph.

It is noticeable in Figs. 2, 3, and 4, that the slip bands on either side of a crystal boundary appear to have no relation

¹ Ewing and Rosenhain, "The Crystalline Structure of Metals," *Philosophical Transactions of the Royal Society*, 1900.

to one another, and are never continued from one side to the other; to such a rule there may be found, however, an important exception. Occasionally upon the strained surface slip bands are seen which do cross a boundary, but in such cases it would seem to be the invariable rule that they do not lose their direction in so crossing, and that their width and intensity are but little affected as they approach the boundary on either side. The actual line of the boundary becomes slightly waved in harmony with the slips, and there would appear to be little tendency for ruffling to take place. Fig. 5, Plate XI., illustrates such a case, and it will be seen that the slip bands cross the boundary on the right of the photograph without being affected by it to any great extent; the absence of ruffling has enabled the boundary itself to be brought into focus with the greater part of the crystals on either side of it.

The theory of the formation of the intercrystalline junctions advanced above would appear to offer a simple explanation of such cases. The fact that where the slip bands cross a boundary they are parallel on either crystal suggests that two crystals have a common gliding plane, and that therefore in this plane their orientations are identical; there would therefore be no displacement of the molecules between the crystals necessary in this plane, and the property of gliding would be continuous from one to the other. In this connection it may be noticed that Ewing and Rosenhain¹ found that slip in iron occurs most readily along the octohedral planes, and, since these are not planes of symmetry, the fact that two crystals possess one in common does not imply identical orientation in every plane. It would be only on the common gliding plane therefore that slips would be transmitted across the boundary, and in all other planes the usual displacements in the orientations which prevent slip would be present.

A particular form of brittleness which is occasionally found in annealed sheet iron has been described by Stead,² and consists in a tendency to break in two directions approximately at 45° to that in which the sheets were rolled. In

¹ *Loc. cit.*

² Stead, *Journal of the Iron and Steel Institute*, 1898, No. 11.

studying the question it was found that such sheets exhibited a marked uniformity in the orientations of their crystals, and the brittleness was ascribed to the consequent parallelism of the cleavage planes. The theory now advanced shows how greatly such uniformity in orientation would tend towards brittleness, since the disturbances in the structure at the crystalline boundaries would be a minimum and the cleavages would be practically continuous from one crystal to another with no amorphous metal to hinder their transmission.

In a previous report¹ the author has shown that the usual and normal adhesion existing between the crystals of a specimen of Swedish iron may be to a very great extent destroyed by heat treatment under certain conditions. The study of slip bands upon an iron so treated reveals, as would be expected, very marked differences in their behaviour from those found upon the normal metal. Fig. 6 by 300 diameters and Fig. 7 by 1000 diameters are photographs taken from the strained surface of strips of Swedish iron from the same sheet as that used in the previous experiments, but in this case the material had previously been subjected to prolonged annealing *in vacuo*. After such treatment the material exhibits a marked inter-crystalline weakness, and fracture invariably takes place along the boundaries. Figs. 6 and 7 were taken from a specimen after straining in tension to a point at which such fracture had just commenced to occur. Although, judged by the number and intensity of the slip bands, the amount of strain produced was approximately the same as that given to the specimens illustrated in Figs. 2 to 5, marked differences are noticeable between the two cases. The ruffling of the surface was very slight, and large areas could be observed under the microscope in uniform sharp focus. The slip bands are of much greater regularity, and do not show anything like the tendency to bend and become finer as they approach the boundaries. It is evident also that they are often continued from one crystal to another without being in any way parallel on either side of the boundary. These observations clearly demonstrate that the peculiarities in the configuration of the slip bands near the boundaries in the normal

¹ *Loc. cit.*

material are due to the strength and resistance which the intercrystalline junctions offer to plastic deformation.

The behaviour of slip bands at or near the intercrystalline junctions has also been studied in the case of other metals, and precisely similar results obtained. It is therefore unnecessary to illustrate them here, but an examination of some of the photographs in Ewing and Rosenhain's original paper¹ will confirm the statement. All the observations demonstrate the resistance to plastic deformation by slip which occurs at the crystal junctions, and suggest the idea that each crystal in a mass of metal is surrounded by and firmly attached to a continuous skin, which, while hard and plastically undeformable, is yet thin and capable of elastic bending and stretching. The attachment of this skin to the crystals is such that its properties are transmitted to the crystalline material near it.

This view and its bearing upon the mechanical properties of metals may now be discussed; but before doing so a comparison may be made as to how the present theory of the formation and structure of the amorphous cement differs in its bearing upon these points from that previously advanced by Rosenhain and Ewen.² The author, apart from his deductions from the theories of crystal structure which have been quoted in Part I., regards two experimental facts as offering a confirmation of his theory, viz.: (1) the continuation of slip from one crystal to another when they possess a common gliding plane, and (2) the marked resistance which the boundaries offer to slip in all cases where such a common gliding plane is absent. The latter observation appears to necessitate a disturbance from the crystalline structure at the boundaries which must extend to far more than molecular dimensions, and would require the "crystal unit," such as that postulated by Rosenhain and Ewen, to be of such a size as to be almost within the resolving powers of the microscope. The following deductions, however, are drawn especially from the microscopical observations, and are independent of any theory as to the precise structure and formation of the intercrystalline medium, provided only it be imagined capable of exerting an influence upon the

¹ *Loc. cit.*

² *Loc. cit.*

deformation by slip of the crystalline material near it of a nature which these observations indicate.

(3) THE MECHANICAL PROPERTIES OF METAL AFTER OVERSTRAIN.

Deformation by slip, which takes place within the crystalline grains of a metal when overstrained, consists in a number of discontinuous movements which, by the configuration of the slip bands formed upon the surface, are shown to occur in a finite number of planes within each crystal. The observations described above demonstrate that motion in any particular gliding plane becomes increasingly more difficult as the crystal boundary is approached, and that in consequence the movement is transferred to other planes lying more nearly parallel to the boundary, and from these distributed and spread out so as to become more continuous. Measurements made upon any particular crystal before and after plastic deformation show that its outline becomes considerably distorted as the strain increases, but since the discontinuous movements of slip upon a limited number of gliding planes which occur in the interior of each crystal, are, by stepping and branching, transferred to an increasingly greater number of planes as the boundary is approached, we must consider that before the actual amorphous skin is reached the distortions become quite continuous, and that the skin itself and the crystalline material near it are only strained by elastic bending. The irregular and polyhedric form of the crystals would allow of a considerable degree of such distortions taking place without necessitating either a change in the volume of a crystal or alterations in the superficial area of its surfaces inconsistent with the elastic properties of the material. If the deformation be continued, a limit to these elastic strains in the amorphous crystal skins will ultimately be reached at certain points, and since the skins are practically incapable of plastic deformation they will break across and the increase of stress thus thrown upon the crystalline matter will result in the rapid fracture of the specimen. The amount of local contraction which occurs before a tensile specimen fractures

will thus depend upon the maximum distortion which the amorphous skins can undergo. Tests made at temperatures at which the amorphous phase becomes considerably more plastic, while still possessing a strength comparable with the crystalline phase, should result in a drawing down of the test-pieces almost to points, and such results have been obtained by the author during his work with Rosenhain referred to above, and in previous high temperature experiments by other workers.

The stress at which a metal ceases to be elastic and undergoes permanent strain is known from the work of Muir¹ and others to be especially dependent upon its previous mechanical and thermal history. Thus, if a metal be plastically strained in tension and the external load removed, it is found that while it remains for a time in a completely plastic state—even under small stresses—yet after a certain lapse of time (which is greatly shortened by slightly raising the temperature) it not only recovers its elasticity but that the elastic limit in tension has been definitely raised. The temporary condition of plasticity has been ascribed to the temporary formation of Beilby's mobile phase between the slip planes, while the subsequent hardening of the metal has been ascribed both to the transformation of the mobile to the amorphous phase² and also to the stopping of "favoured slip planes."³ While both these causes bring about a hardening of the metal, which in cases of severe deformation such as those produced by rolling, hammering, and wire-drawing, may overshadow the hardening produced by any other cause, yet they fail by themselves to provide an explanation of some further results which are prominent after such overstrain, as may be applied in a tensile testing machine.

A metal has not only one, but two elastic limits. In a specimen of metal in a perfectly annealed condition, these two elastic limits lie equally upon either side of the zero of stress—that is, a limit in tension and another in compression. As de-

¹ Muir, "On the Recovery of Iron from Overstrain," *Philosophical Transactions of the Royal Society*, 1900, vol. excii. p. 1.

² Beilby, "The Hard and Soft State in Metals," *Philosophical Magazine*, August 1904.

³ Rosenhain, "The Deformation and Fracture of Iron and Steel," *Journal of the Iron and Steel Institute*, 1906, No. II.



FIG. 1.
Magnified 300 diameters.



FIG. 2.
Magnified 300 diameters.



FIG. 3.
Magnified 300 diameters.

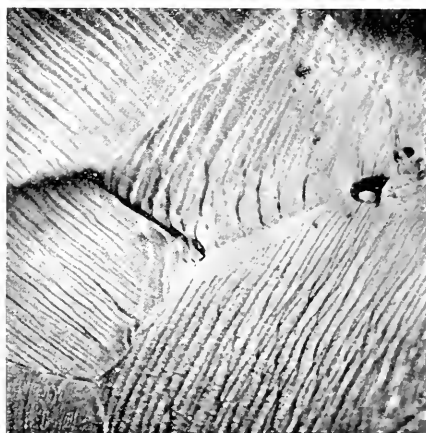


FIG. 4.
Magnified 300 diameters.

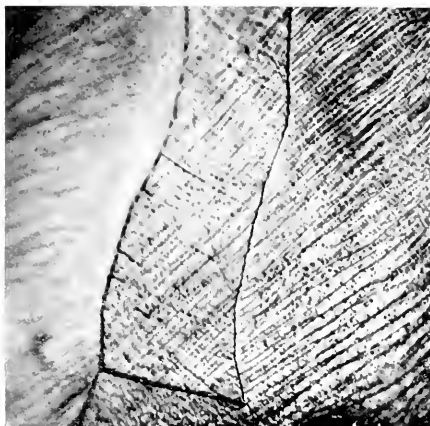


FIG. 5.
Magnified 300 diameters.



FIG. 6.
Magnified 300 diameters.

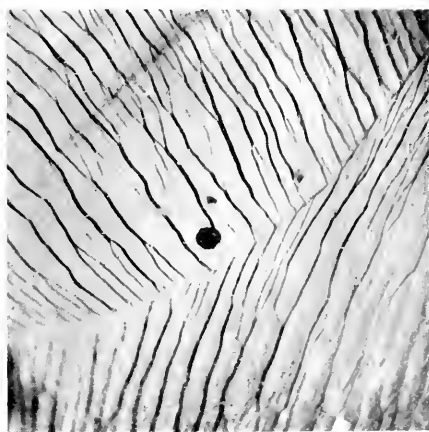


FIG. 7.
Magnified 300 diameters.

scribed above, the elastic limit in tension may be artificially raised by overstrain in tension and subsequent recovery, but in doing so it is found that the elastic range of stress in compression is at the same time reduced. Similarly, increasing the elastic range in compression by overstrain in compression at the same time reduces the elastic range in tension. By sufficient overstrain in either direction it is in fact possible to move one elastic limit right up to the zero of stress, so that the material behaves entirely plastically under stress in one direction.¹

The finite discontinuous movements of slip by which the plastic deformation of the crystals of a metal takes place have been shown by the observations described above to extend towards the boundaries by smaller and more numerous movements which finally become continuous and elastic as the skin is reached. The elastic strains so produced would only be completely relieved on removal of the load if the movements of slip were at the same time reversed. Both from the permanent deformation of the specimen and from the permanent appearance of slip bands, we know that the motions are not so reversed, and therefore the elastic strains must remain and cause permanent internal stresses in the specimen. The state of affairs can best be conceived by imagining that after overstrain each crystal is enclosed and held within a *stretched* skin holding it so as to oppose further strain in the same direction. Subsequent stresses applied to the specimen in the same direction as those causing the original overstrain will now have to overcome the internal stresses in the crystal skins as well as the normal resistance to slip of the material, and hence the elastic limit in this direction will be raised. Stresses applied in the opposite direction will, however, be assisted by the internal stresses, and hence the elastic limit in this direction will be lowered.

A simple mechanical model may be used to illustrate the theory. In Fig. 8 (*a*) A and B represent two bodies in contact and capable of sliding one over the other; such sliding being in part resisted by friction and in part by a spring C which also holds them in contact. If forces are applied to

¹ Bairstow, "The Elastic Limit of Iron and Steel under Cyclical Variations of Stress." *Philosophical Transactions of the Royal Society, A*, vol. ccx. pp. 35-55.

the bodies tending to slide them over one another, no motion will ensue until the frictional resistance has been overcome; but directly this is exceeded, the one will commence to slide over the other, the amount of actual movement being controlled by the amount by which the excess of the forces over that required to overcome frictional resistance can stretch the spring C. Suppose a certain movement takes place so that the position is as shown in Fig. 8 (b). If the external forces are now removed, and the tension remaining in the spring is insufficient to overcome the frictional resistance, no slipping back will take place, and the spring will remain permanently stretched. On again applying forces in the same directions as before, the tension remaining in the spring, as well as the frictional resistance, must be overcome before motion takes

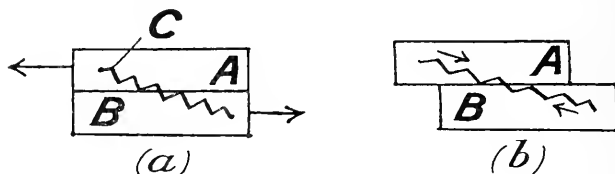


FIG. 8.—Magnified 300 diameters.

place. If, however, forces are applied in the opposite directions the tension of the spring will assist them in overcoming the frictional resistance. The bodies A and B represent two portions of crystals on either side of a gliding plane and the spring C the amorphous cement enclosing the crystal. The spring has been so placed in the model that any sliding produces a greater pressure between the two bodies, and hence an increase in the frictional resistance, and corresponds to the total surrounding of the crystal by amorphous matter.

The analogy of this model may be pushed further to provide an explanation of the yield point. It is known that static friction is always greater than dynamic, and that when once motion takes place between two surfaces the frictional resistance is reduced. In such a model, therefore, directly sliding begins it will proceed suddenly without the external

forces being increased, until the force equalling the difference between the resistance to dynamic and static friction is taken up in the spring.

The author had hoped to test the above theory by some further experimental work. Since the effect of overstrain upon the elastic limits has been ascribed to the resistance which the material at or near the intercrystalline junctions offers to discontinuous slips, and since it has been shown that this resistance may be to a very great extent destroyed in Swedish iron by suitable treatment, it would follow that such intercrystallinely weak material should behave in a very abnormal fashion as regards the alterations in its elastic limits when tested after overstrain. Experimental difficulties have, however, been encountered, which up to the present have prevented the author from producing such intercrystalline weakness in test-pieces of suitable size for the delicate extensometer measurements required. He has, therefore, been compelled to defer these experiments to some future occasion.

The microscopic work of the present research has been carried out in the Department of Metallurgy and Metallurgical Chemistry of the National Physical Laboratory, and the author would like to express his thanks to the authorities and especially to Dr. R. T. Glazebrook, C.B., F.R.S., the Director, and to Dr. W. Rosenhain for their permission to use the facilities of the Laboratory. The author has had many opportunities of discussing the work with Dr. Rosenhain, and although their views are not in complete agreement as to detail, yet he would like to offer his sincerest thanks for the assistance which these discussions have given him.

A STUDY OF THE CONSTITUTION OF CARBON-MOLYBDENUM STEELS

WITH AN APPENDIX ON
THE MECHANICAL PROPERTIES OF SOME LOW
MOLYBDENUM ALLOY STEELS.

BY THOMAS SWINDEN, M.MET. (SHEFFIELD).¹

INTRODUCTION.

IN a previous report on this subject² the author dealt with the mechanical properties, electrical resistivity, carbide analysis, and micrographic analyses of a series of carbon-molybdenum steels in various conditions of treatment. The object of the present research is further to examine the properties of these steels in order to obtain more exact knowledge of their constitution.

One of the most interesting facts in regard to molybdenum steels is the lowering of the critical points under certain conditions. The tungsten steels, which behave in a somewhat analogous manner, have been exhaustively examined by the author,³ and similar work has also been done by Professor Carpenter and Mr. Edwards with high-speed steels. Opinion is divided as to the nature of the constitution, and the author has therefore made a thorough examination of the conditions affecting the lowering and recovery of the critical points. The microstructure revealed in specimens after various heat treatments has been examined, and the electrical resistivity, a valuable asset in the study of chemical constitution, has been determined.

¹ Received April 1, 1913.

² *Carnegie Scholarship Memoirs*, vol. iii. (1911).

³ "Carbon-tungsten Steels," *Journal of the Iron and Steel Institute*, 1907, No. I. pp. 291-324; "The Constitution of Carbon-tungsten Steels," *Ibid.*, 1909, No. II.

The results will be presented in the following order:—

Recalescence curves.

Electrical resistivity of quenched specimens.

Microstructure of quenched specimens.

Hardness tests of quenched specimens.

General discussion of the results and their bearing on the constitution of molybdenum steel.

STEELS EMPLOYED.

The chemical analyses of the steels used are given in Table I., and details of the manufacture are to be found in the earlier report (*loc. cit.*).

RECALESCENCE CURVES.

Description of Apparatus.—These curves have been taken on a direct-reading installation designed by the author for the new physical laboratory of Messrs S. Fox & Co., Limited, Stocksbridge Works. The installation is of a similar type to that installed by Professor Arnold in the Sheffield University and described by him.¹

A thermo-couple placed in close contact with the specimen is connected up to a dead beat reflecting galvanometer, and the spot of light is reflected on a metre scale. The operator signals on a chronograph recorder every time a hair line in the reflected spot is coincident with a millimetre mark on the scale. The recorder tape then shows the number of seconds per millimetre fall (or rise); that is, the inverse of the rate of cooling (or heating); and by plotting this as an ordinate with actual temperature as abscissa, any arrest in the rate of heating or cooling is very clearly shown.

A plan of the apparatus is shown in Fig. 1.

The essential parts are all mounted on brick pillars to avoid vibration, as shown dotted in the sketch. The galvanometer, G, is of the Ayrton-Mather pattern, and has a resistance of 21·7

¹ *Proceedings of the Institution of Civil Engineers*, vol. cxxiii., Part I. pp. 127–162.

TABLE I.—*Analyses of Steels.*

Mark.	Carbon per Cent.	Molybdenum per Cent.	Manganese per Cent.	Silicon per Cent.	Sulphur per Cent.	Phosphorus per Cent.	Arsenic per Cent.	Nitrogen per Cent.	Aluminium per Cent.
1	0.195	1.030	0.218	0.047	0.025	0.016	0.018	0.0060	0.005
2	0.445	1.054	0.230	0.087	0.026	0.014	0.016	0.0056	...
3	0.869	1.018	0.270	0.122	0.039	0.018	0.019	0.0055	...
4	1.215	1.096	0.250	0.124	0.032	0.017	0.017	0.0043	...
5	0.246	2.176	0.216	0.064	0.026	0.015	0.020	0.0004	...
6	0.442	2.181	0.270	0.075	0.025	0.015	0.020	0.0038	...
7	0.883	2.186	0.263	0.078	0.025	0.014	0.019	0.0066	...
8	1.210	2.109	0.238	0.071	0.023	0.015	0.017	0.0053	...
9	0.190	4.110	0.242	0.036	0.020	0.017	0.019	0.0018	...
10	0.487	4.009	0.292	0.038	0.024	0.016	0.022	0.0028	...
11	0.865	4.002	0.230	0.051	0.022	0.016	0.021	0.0033	...
12	1.060	4.019	0.230	0.039	0.020	0.016	0.018	0.0056	...
13	0.135	8.012	0.216	0.049	0.022	0.017	0.021	0.0034	0.004
14	0.361	8.167	0.274	0.056	0.025	0.016	0.011	0.0048	...
15	0.445	8.109	0.252	0.080	0.022	0.019	0.019	0.0037	...
16	0.775	7.847	0.230	0.041	0.024	0.017	0.022	0.0037	...
17	1.125	7.920	0.234	0.075	0.027	0.017	0.020	0.0043	...
0	1.360	2.540	0.205	0.122	0.024	0.022	0.016	0.0058	...

ohms with a damping coil of 75 ohms. The deflection for 1 microvolt at 2 metres, the working distance of the mirror, is 3.71 millimetres. The zero creep for full deflection of 1 metre for 1 hour is 3 millimetres. The galvanometer is practically dead beat, and there has been no difficulty whatever due to unsteadiness of the galvanometer, as the curves will testify.

A 4-foot platinum-platinum rhodium couple was used, connected up by Peakes compensating leads to the galvanometer. By means of these leads, which are designed to have the same electromotive force as the couple, the cold junction can be brought to the galvanometer terminals. The junction is kept protected from irregular heating by a water bath.

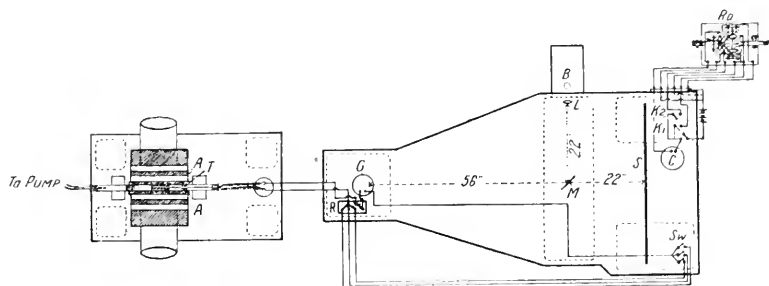


FIG. 1.—Plan of Recalescence Apparatus.

In order to increase the usefulness of the apparatus a special series of resistances, R , have been introduced into the circuit. These are connected to switches, Sw , on the operating table, and by switching in one or other of the resistances the metre scale is made to represent a range of 600° , 1000° , or 1500° C. The cold junction temperature is read off on a thermometer placed on this resistance box. It varied during the period of the work between 14° and 20° , and half this value has been added to the observed scale reading in stating the temperatures.

The light is supplied by an incandescent burner, B , standing outside the hood which envelops the rest of the apparatus. The light is brought into a parallel beam by lens, L , which carries a very fine pointer. A plane mirror, M , placed at 45°

with a slight upward tilt reflects on to the galvanometer mirror, which in turn reflects back on to a 1-metre translucent scale, S. The distance from the pointer to galvanometer mirror and from the latter to the scale are each 2 metres, the focal length of the mirror.

The specimens were heated in a gas muffle furnace of special design, shown in horizontal section in Fig. 1. Five powerful burners play on to a 2-inch diameter silica tube, T, and the gases are then led down again behind protecting arches, A, to two side chimneys. The length of the silica protecting tube was 12 inches. The furnace is worthy of note, as a perfectly even temperature could be obtained within the inner 9 inches even at low heats, whilst a maximum of 1200°C . in the inner tube could be reached without blast in about $1\frac{1}{2}$ hours. The inner tube carrying the specimen was of silica, 24 inches by $\frac{7}{8}$ inch internal, and the annular space between the two tubes was packed fairly tightly with asbestos fibre, to avoid irregular heating or cooling.

The heatings were done *in vacuo*, the air being pumped out by means of a strong Kortings water pump. A vacuum of $\frac{1}{2}$ inch could be obtained in a few minutes, and the tube then sealed off by a glass stop-cock. As the vacuum fell due to heating (and shown on a small mercury gauge attached), the pump was started, the tap opened, and vacuum restored. The pieces were perfectly bright and silvery even after twenty heatings, some reaching 1200°C .

The chronograph recorder, R, is connected through a clock, C, making half-second contacts, and marks off the half-second intervals by pin-pricks in the tape. The tapping keys, K1 and K2, actuate the left and right signals respectively. A 6-volt accumulator serves the recorder system.

The apparatus has worked in a perfectly satisfactory manner, and duplicate curves taken after an interval of six months' continuous work entirely confirm the constancy of the apparatus.

The standardisation was carried out in the usual way with boiling water, solidifying tin, lead, zinc, sodium chloride, and melting gold (wire) as standard temperatures.

The specimens were $\frac{1}{2}$ inch diameter, 2 inches long, drilled

$\frac{3}{4}$ inch with a $\frac{1}{4}$ -inch hole and recessed for a further $\frac{1}{4}$ inch with a $\frac{1}{16}$ -inch hole. The couple was in this latter recess, and was therefore almost in contact with the steel. The specimens had a mass of 80 grammes. The calibration on the 1500° scale as used for all these curves corresponded to $2\cdot0^{\circ}$ C. per millimetre at 300° and $1\cdot2^{\circ}$ per millimetre at 1000° .

Scheme of Experiment Followed.—The experiments are designed to investigate the influence of the following conditions upon the position of the critical points in heating and cooling.

(A) *The Lowering of the Point.*

1. Rate of heating.
2. Maximum temperature.
3. Time of soaking at maximum temperature.
4. Rate of cooling.
5. Carbon content with constant molybdenum.
6. Molybdenum content with constant carbon.

(B) *The Recovery of a Normal Point after Lowering has taken place.*

1. Rate of reheating.
2. Time of soaking.
3. Exact previous treatment.

In cases where the specimen was "soaked" for a certain period, the variation in temperature did not exceed 5° C. either side of that stated, for temperatures up to 1000° . For 1100° and 1200° the temperature was kept within 10° either side.

The curves are set out in Figs. 2–18 and Plates XII. and XIII., and will be dealt with as illustrating the questions above enumerated.

DESCRIPTION OF COOLING CURVES.

(A) CONDITIONS AFFECTING THE LOWERING OF THE POINT.

1. *Rate of Heating.*

In this case, as in all others under the above heading, it is assumed that we start with a piece showing normal points in

a curve taken from below the lowering temperature. Special experiments with Steel 4 show that a slow heating up and two hours' soaking at 795° gives a point identical with that produced by quick heating to the same temperature and no soaking (curves 4B1 and 4A1). Also by slightly exceeding the "lowering temperature" as in Curves 4B2 and 4B3, the rate of heating through the points being widely different as seen by the heating curves, the resulting cooling curves show identical points. Curves 10, 11 on Steel 7 also confirm this statement.

It may therefore be assumed that, within the range of these experiments, the rate of heating has no influence on the position of the change point. This statement is not quite true in regard to the recovery of the lowered point, as will be shown later.

2. *Influence of Maximum Temperature.*

This is of paramount importance in its influence on the position of the change points. It has previously been shown by the author (*loc. cit.*) that for tungsten steels there exists a certain definite initial temperature of heating known as the "lowering temperature," which marks the commencement of a lowering action on the change points. If heated to anywhere below this temperature, the curves are identical with those obtained from carbon steels of similar carbon content.

The scheme for examining this phenomenon in molybdenum steels has been to heat at as near a uniform rate as possible—about 18° C. per minute—soak for half an hour at temperatures of 800° , 850° , 900° , 950° , 1000° , 1100° , and 1200° , and take cooling curves.

The "lowering temperature" figures quoted should therefore be taken to apply to these conditions. In certain cases other conditions have been employed to examine other specific details governing the lowering.

The simplest method of dealing with the results will be to take the steels in batches of similar molybdenum content and increasing carbon.

Steel 1: Carbon 0.20, Molybdenum 1.03 (Fig. 2).—The curves from 800° only show a small point about 770° , probably Ar₂,

and irregularities from 715° down to 675° . No very clear

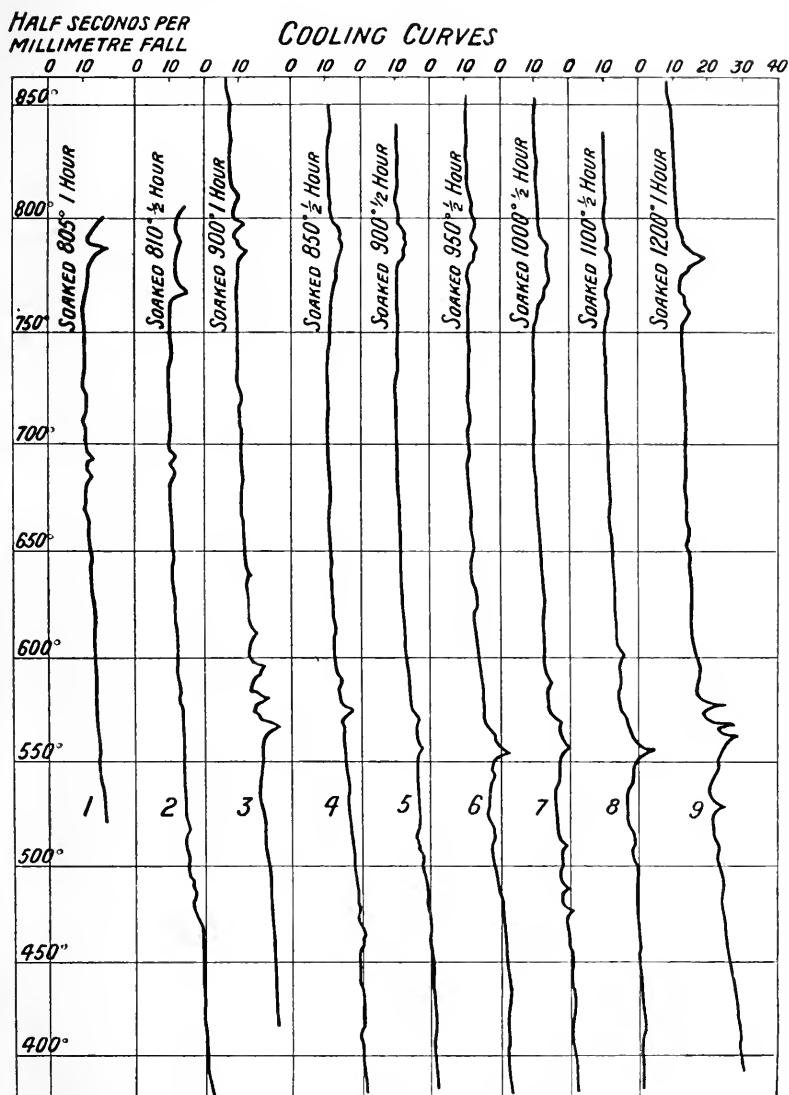


FIG. 2.—Steel No. 1. Carbon, 0.20 per Cent. ; Molybdenum, 1.03 per Cent.

point is found at the normal temperature for Ar, although Curve 2 confirms a small point at about 700° .

Cooled from 850° or beyond, distinct signs of a broad recalescence range is seen about 570° . The later curves do not show a definite Ar₃, but the small point found in all the curves at about 760° to 770° is most probably Ar₂.

The original Ar₁ probably exists about 700° , and is first lowered by a temperature between 810° and 850° . Ar₂ is

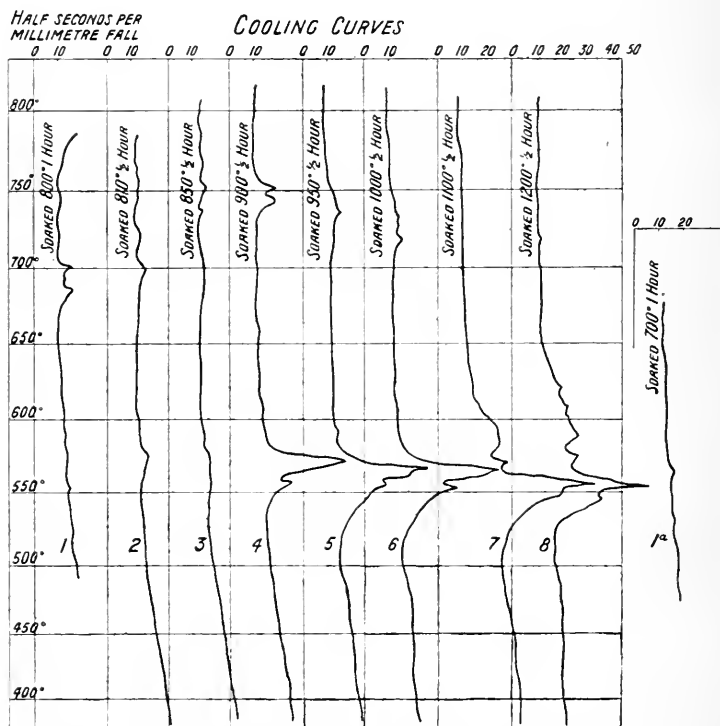


FIG. 3.—Steel No. 2. Carbon, 0.45 per Cent.; Molybdenum, 1.05 per Cent.

practically constant throughout at about 770° . The lowered point is about 560° , and it is interesting to note that this temperature, reached by cooling from 850° , is confirmed by cooling from 1200° .

Steel 2: Carbon 0.45, Molybdenum 1.05 (Fig. 3).—The curve from 800° gives a comparatively small broken point at 690° , and Curve 2 from 810° has partly lowered this. Curve 4

from 900° gives a very distinct point at 570° , and also a double-headed point at 750° corresponding to Ar3.2. Further heating strengthens the low point at 560° , and gradually obliterates Ar3.2 in its normal position. The latter probably forms the upper portion of the low recalescence in Curve 8.

The final curve on this steel, taken after soaking at 700° , shows that there is no recalescence on cooling if the normal AC points have not been passed in heating.

Steel 3: Carbon 0.87, Molybdenum 1.02 (Plate XII.).—This steel has been selected for a special investigation into the conditions affecting lowering and recovery. The question of recovery will be dealt with later under a special section. A large normal Ar3.2.1 point is seen at 713° after cooling from 800° , but this is soon displaced. The curves on this steel illustrate very clearly the mobility of the recalescence temperature between the normal and the low points by altering the initial temperature.

Curve 7 from 906° gives the low point almost down to its limit, a fact of interest when considered with the high molybdenum steels. The lowest point obtained is at 556° , and is practically constant when the initial temperature has exceeded 1000° .

Steel 4: Carbon 1.22, Molybdenum 1.09 (Fig. 4).—This series is of exceptional interest as showing the dominating influence of initial temperature on the character of the change points, and through these of course on the constitution of the alloy.

Curves 1 and 2, afterwards confirmed by 4A1, 4A2, and 4B1, show that if the initial temperature does not exceed 850° the normal point Ar3.2.1 is constant at 715° . When the temperature is taken to 900° two points are found at 685° and 585° respectively. It was thought that perhaps the two points were due to too rapid heating or insufficient soaking, but Curves 4B2 and 4B3 entirely refute this argument, and go to prove that a change is taking place at two definite points.

The point assumes practically a simple shape after cooling from 950° , and is almost at the lowest limit. Increased initial temperature up to 1200° lowers the point slightly, giving a final value of 562° for the maximum.

Steel 5: Carbon 0.25, Molybdenum 2.18 (Fig. 5).—Curve 1 from 805° shows only a small evolution about 750° —probably Ar2—and a somewhat elongation point about 690° —probably Ar1. The second curve from 810° has further dispersed the Ar1 point, and no very definite point is found until Curve 4 from 900° . The Ar1 point is then seen distinctly, lowered to about 540° . The 950° curve strengthens this, and also still

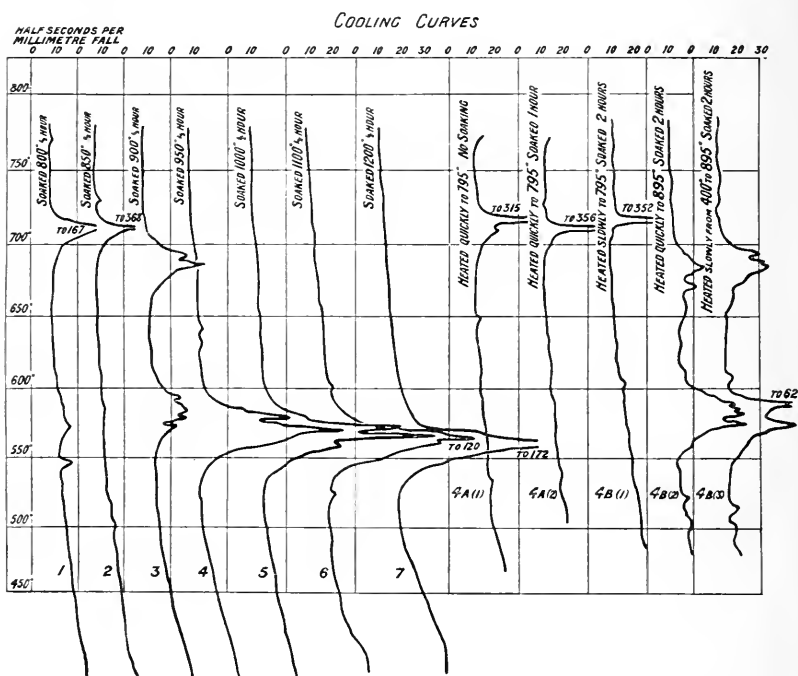


FIG. 4.—Steel No. 4. Carbon, 1.21 per Cent.; Molybdenum, 1.10 per Cent.

shows a small recalescence about 750° —probably Ar2 or Ar3.2. Finally from 1200° irregularities appear from about 570° to 470° , with a maximum at 515° . The broadened upper limit probably corresponds to the lowered Ar3.2 points.

Steel 6: Carbon 0.44, Molybdenum 2.18 (Fig. 6).—Cooling from 810° shows a small point about 770° with a distinct wide point at 685° . The usual soaking at 850° lowers Ar1 almost to its lowest limit with a maximum at 540° . The lowering

temperature under these conditions is therefore about 830° . Curve 7 from 900° gives a very distinct, wide, low point with maximum at 525° , and also a double-headed point about 750° , no doubt Ar3.2. Further heating obliterates this and widens the low point.

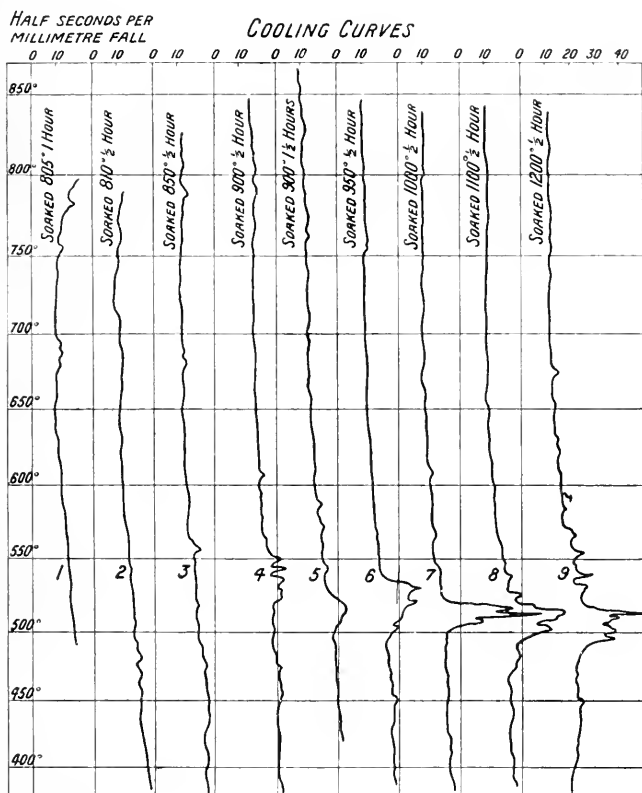


FIG. 5.—Steel No. 5. Carbon, 0.25 per Cent. ; Molybdenum, 2.17 per Cent.

Steel 7: Carbon 0.88, Molybdenum 2.19 (Fig. 7).—This series illustrates in a very clear manner the influence of initial temperature on the position of the critical points. Cooled from 790° , a very large Ar3.2.1 point is found at 720° . This falls gradually as the initial temperature is increased, and it will be noted that over 1000° is required in this case to reach the "final" position, as shown by the 1190° Curve 7. The extreme

sensitivity of the steel to small differences in initial temperature between 800° and 900° is noteworthy.

Steel 8: Carbon 1.21, Molybdenum 2.10 (Fig. 8).—The normal Ar3.2.1 point found in Curve 1 at 712° is dislodged by 850° as a large point, and lowered successively till the initial

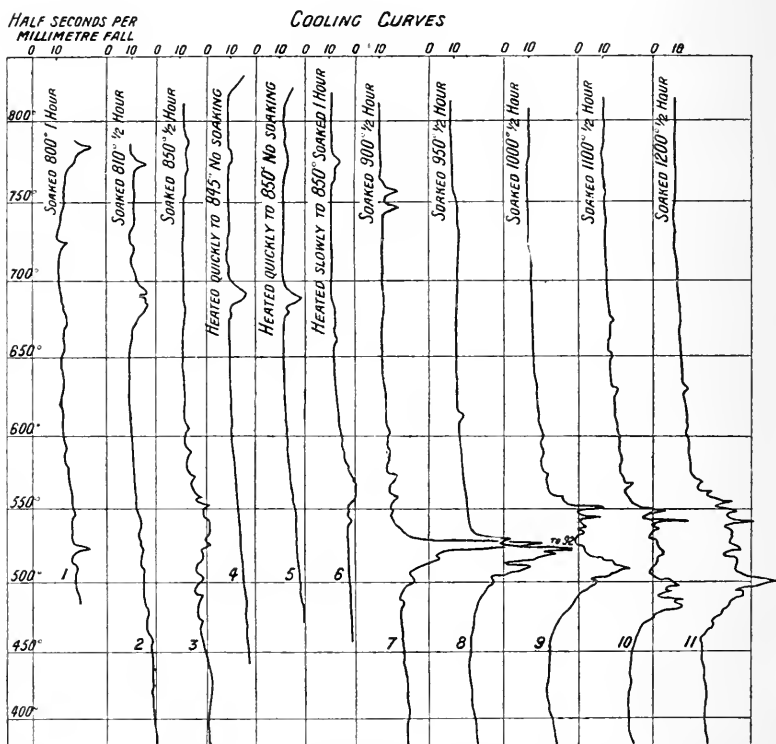


FIG. 6.—Steel No. 6. Carbon, 0.44 per Cent. ; Molybdenum, 2.18 per Cent.

temperature reaches 1100° . The lower point at 530° is confirmed by cooling from 1200° .

Steel 9: Carbon 0.19, Molybdenum 4.11 (Fig. 9).—Cooled from 800° , this steel shows a small point at 760° , probably Ar2, but no distinct Ar1. Curve 2 from 820° fails to show Ar1 in a normal position, but small irregularities are found in the neighbourhood of 530° . No distinct Ar3 is found, but a small recalescence at 750° to 760° is constant throughout, and

is almost certainly Ar₂. The Curve 8 from 1200° produces a distinct broad point with maximum at 475°, with smaller irregularities, possibly due to the lowered Ar₃ showing from 600° downwards.

Steel 10½: Carbon 0.49, Molybdenum 4.01 (Fig. 10).—This gives a remarkably small normal Ar₁ point on cooling from 810°, considering the carbon content. It would appear as though the lowering temperature was actually about 800°.

HALF SECONDS PER
MILLIMETRE FALL

COOLING CURVES

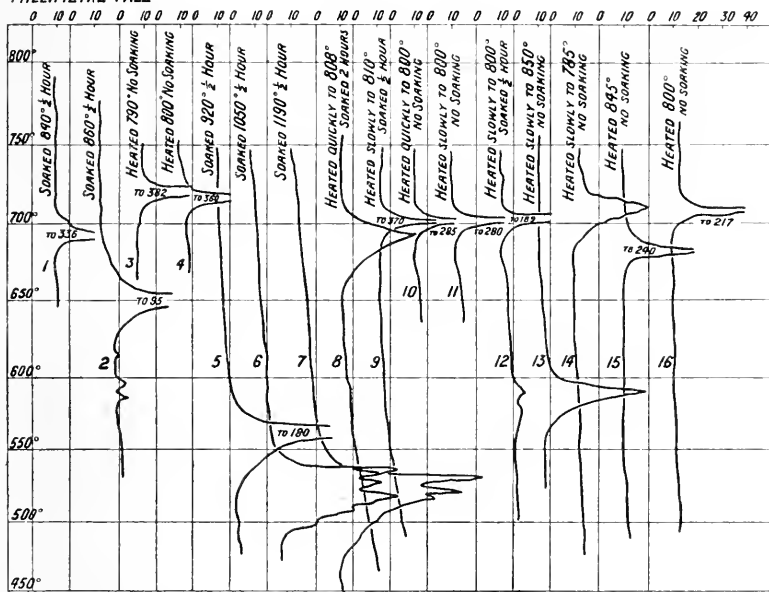


FIG. 7.—Steel No. 7. Carbon, 0.88 per Cent.; Molybdenum, 2.19 per Cent.

since even 810° gives a slight low point. Curve 5 from 850° gives a fairly distinct recalescence about 770°. The low point increases in size, and is lowered to its "limit" by 1100°, when a broad point with maximum about 460° is seen. Curve 10 from 1200° broadens the point without further lowering. The small Ar_{3.2} point is not seen when the initial temperature has exceeded 950°.

Steel 11: Carbon 0.87, Molybdenum 4.00 (Fig. 11).—This series also forms an interesting study in the mobility of the
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points due to variation in initial temperature, &c. Curve 2 from 790° gives a normal Ar1 at 719° , and there is also a small upper point present in all the low temperature curves. Lowering soon occurs as the temperature increases, Curve 5 from 860° giving a clean large point about 550° , but 1200° is necessary to give the lowest position of the point in the form of a broad, irregular point with a maximum about 470° .

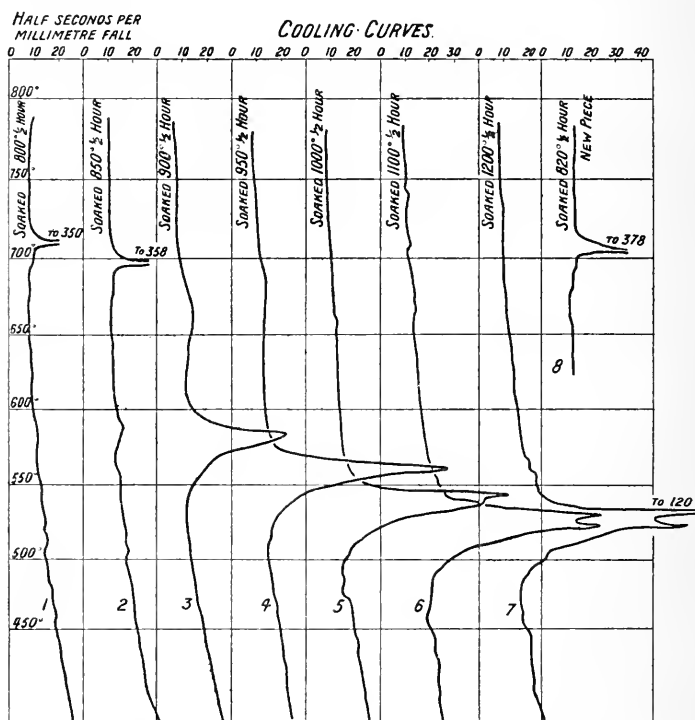


FIG. 8.—Steel No. 8. Carbon, 1.21 per Cent. ; Molybdenum, 2.11 per Cent.

Steel 12: Carbon 1.06, Molybdenum 4.02 (Fig. 12).—The normal Ar3.2.1 shown in the first curve at 692° is gradually lowered as the temperature increases, right up to 1200° . The final curve shows a broad point with maximum about 475° .

Steel 13: Carbon 0.14, Molybdenum 8.01 (Fig. 13).—The curves from 710° , 810° , and 850° give no sign of a normal Ar1. There are faint signs of Ar2 about 760° , and also small

irregularities down about 450° , probably corresponding to a lower Ar1. As the temperature increases Ar2 is still seen faintly, and the low irregularities become a little more definite, but it appears to extend in the final curve from about 600° to 430° .

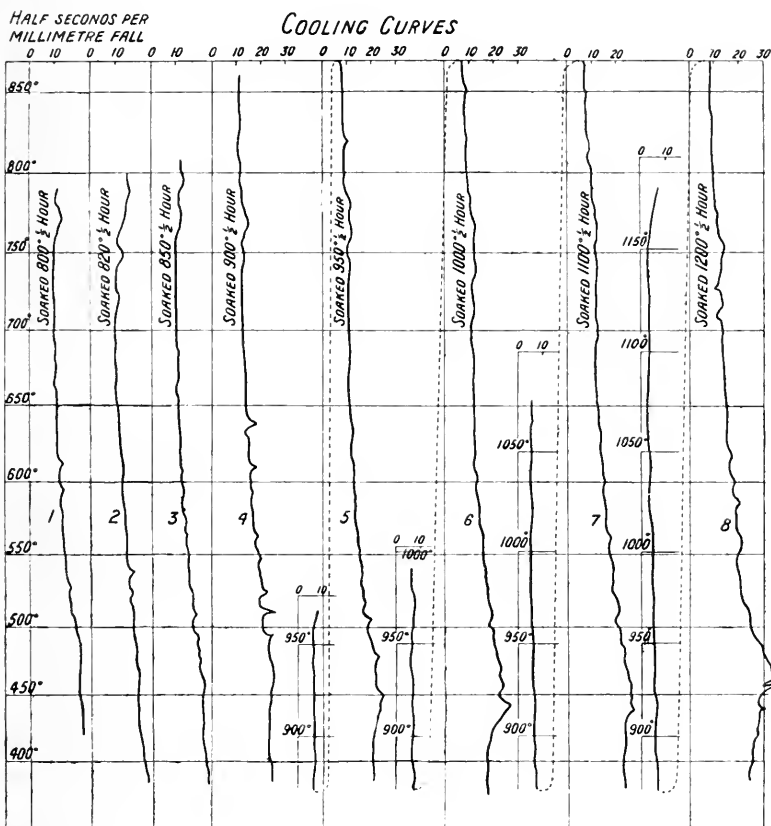


FIG. 9.—Steel No. 9. Carbon, 0.19 per Cent.; Molybdenum, 4.11 per Cent.

Steel 14: Carbon 0.36, Molybdenum 8.17 (Fig. 14).—Note the absence of a normal Ar1 point in the early curves. Even Curve 1 from 810° shows signs of a low recalescence, which are made more distinct in approximately the same position as the temperature increases. In Curves 4 to 7 are signs of a very small Ar2 point about 750° , but even in the curve

from 1200° the low recalescence is a broad point ranging from 600° to 400° .

Steel 15: Carbon 0.45, Molybdenum 8.11 (Fig. 15).—This series is very similar to Steel 14 in that no normal Ar1 point

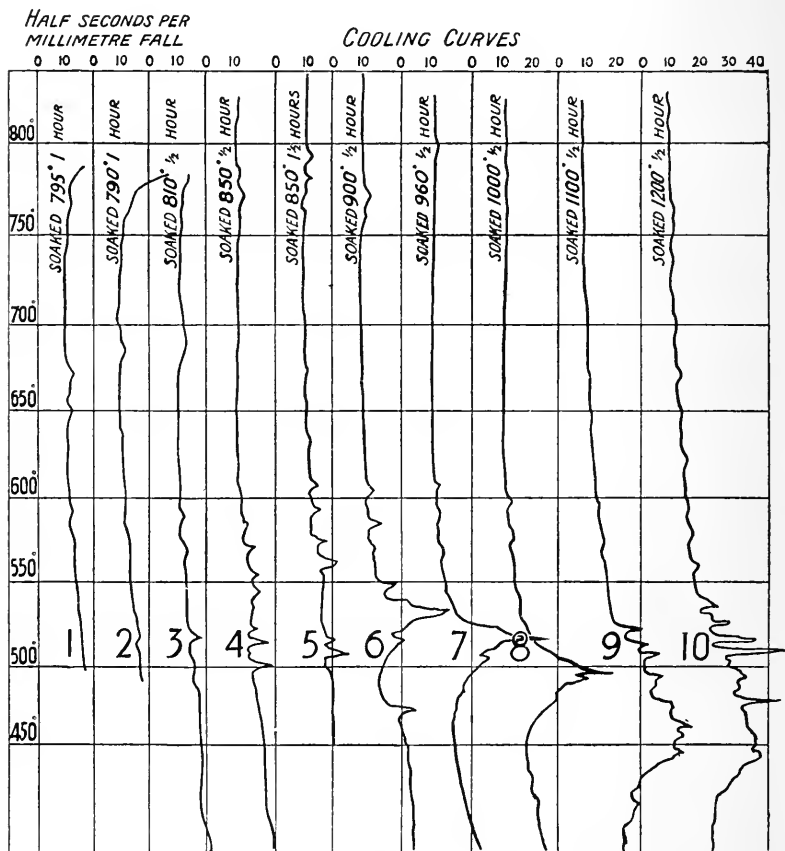


FIG. 10.—Steel No. 10. Carbon, 0.49 per Cent. ; Molybdenum, 4 per Cent.

is found. The distinct pieces of this steel were examined, and every possible care taken to detect a point, but without success. The only conclusion appears to be that the lowering point approaches very nearly to the Ac points in heating. Curve 6 from 1100° gives a distinct point at 450° , but

Curve 7 from 1200° shows a very wide recalescence from 600° downwards, with a maximum about 445°.

Steel 16: Carbon 0.78, Molybdenum 7.85 (Fig. 16).—A sharp, medium-sized point is found in Curve 1 at 703°, but 815° has weakened this, and 850° brings in numerous small points down to 500°. A curious feature in this series is the

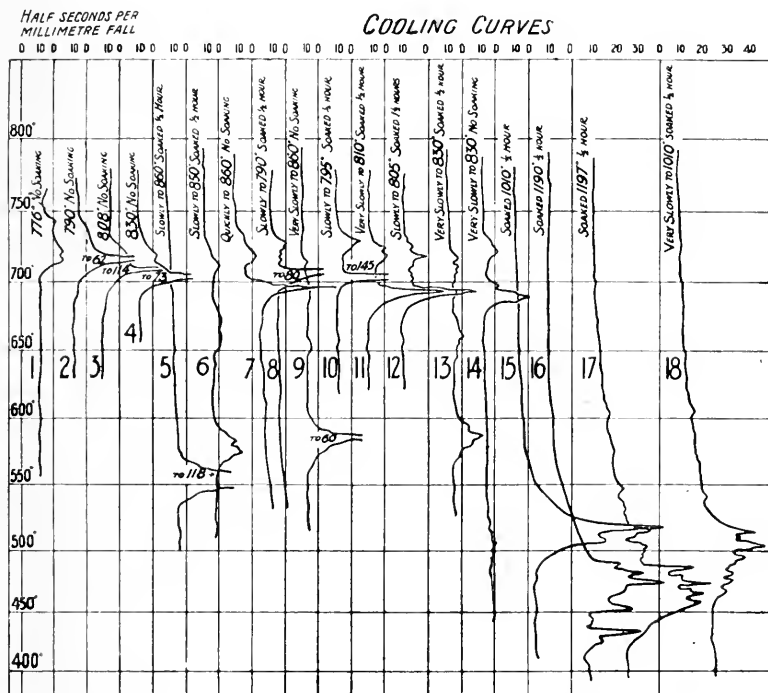


FIG. 11.—Steel No. 11. Carbon, 0.87 per Cent.; Molybdenum, 4 per Cent.

large, crisp point shown at 545° in Curve 5 from 900°, which is gradually dispersed by subsequent heating until 1200° gives a broad, irregular point with maximum about 450°. Note should also be made of the small point showing in Curve 2 at 760°, which is gradually lost as the temperature increases.

Steel 17: Carbon 1.12, Molybdenum 7.92 (Fig. 17).—Curve 1 from 810° shows a particularly large Ar_{3.2.1} point at 692°. Curves 3, 4, and 5 demonstrate very clearly that the position

of the point is dependent on the maximum temperature, and note should be made here again of the large points shown in these curves. The final point in Curve 7 from 1200° is very broken, with a maximum about 430° .

Steel 0: Carbon 1.36, Molybdenum 2.54 (Fig. 18).—A large

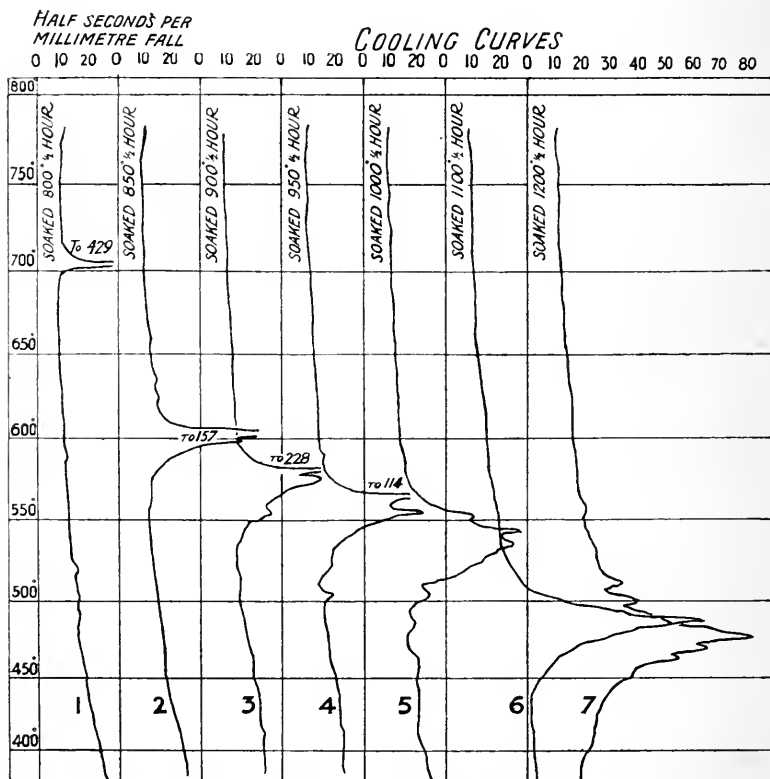


FIG. 12.—Steel No. 12. Carbon, 1.06 per Cent. ; Molybdenum, 4.02 per Cent.

normal point is seen in Curve 1 from 810° to 715° . This set forms a good illustration of the splitting and lowering as the temperature increases. The lowest limit is reached with a temperature of 1100° , the maximum of the final low point showing at 515° .

In summary under this sub-heading several important items should be noted:—

- (a) In steels where Ar₂ exists free, it is not affected by the initial temperature, but where associated with Ar₃ the combined point is lowered after Ar₁. The point Ar₃ is not shown distinctly in any of the steels.

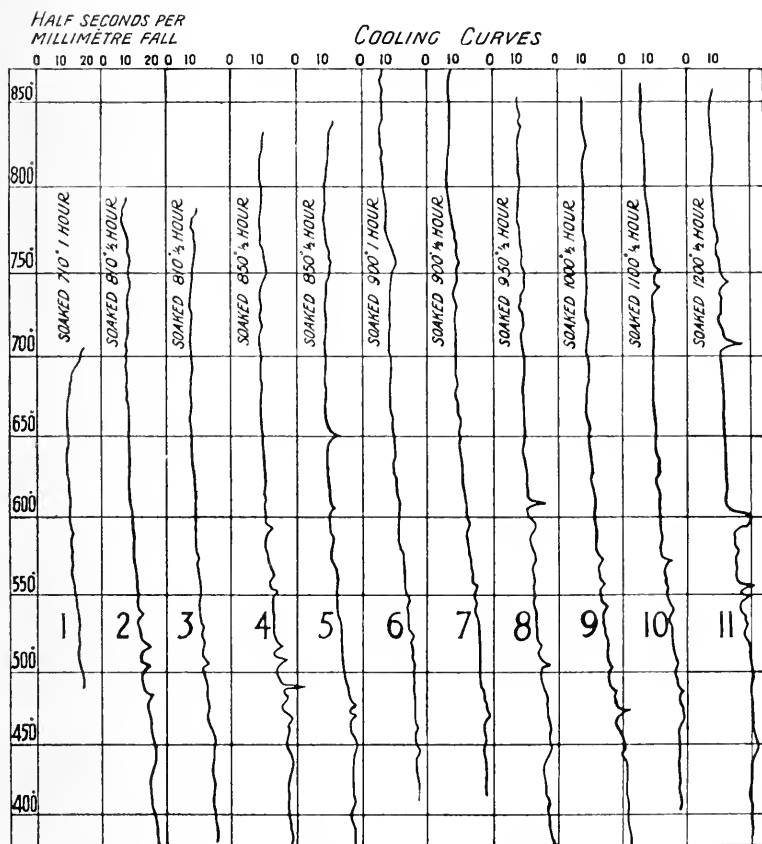


FIG. 13.—Steel No. 13. Carbon, 0.14 per Cent.; Molybdenum, 8.01 per Cent.

- (b) The split or double point shown when lowering has just begun is shown by experiments on Steel 4 to correspond to a definite state of the material, and is not due to rate of heating or insufficient soaking. The curves on this steel also prove the lowering to commence only when the initial temperature has exceeded a certain definite point.

- (c) Although large points are found in the high carbon steel, the 0.20 and 0.45 carbon series show extremely small Ar1 points. Particularly is this so in the high molybdenum series, where no normal Ar1 can be

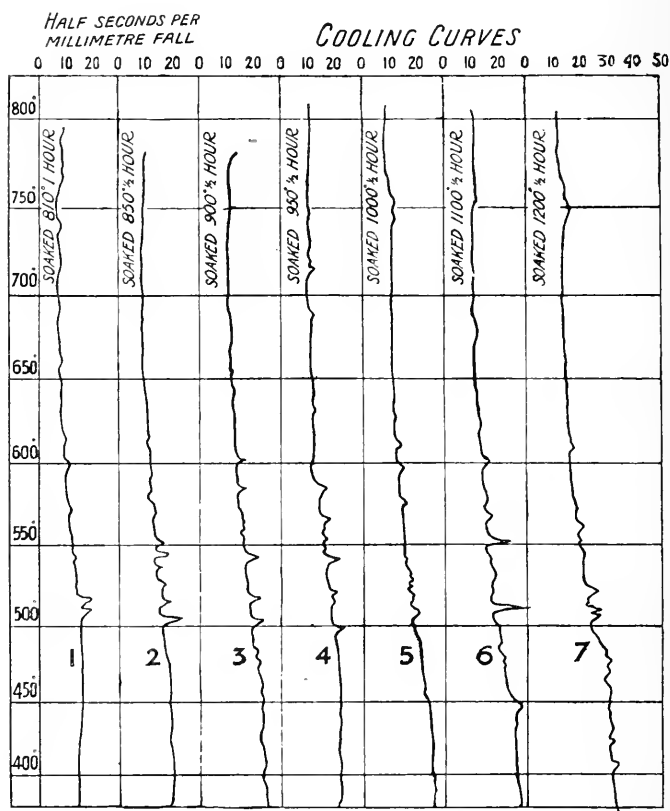


FIG. 14.—Steel No. 14. Carbon, 0.36 per Cent. ; Molybdenum, 8.12 per Cent.

found until the carbon reaches 0.78 per cent. The lowered point is rendered very broad and vague as the molybdenum increases.

- (d) As the molybdenum increases, the carbon content being constant, a higher temperature is required to bring about a complete lowering of the point. Also in any

one series of constant molybdenum a higher temperature is similarly required as the carbon increases.

3. Influence of Time of Soaking.

Curves 4A1 and 4A2 on Steel 4, and also Curves 11 and 12 on Steel 7, show that time of soaking below the lowering

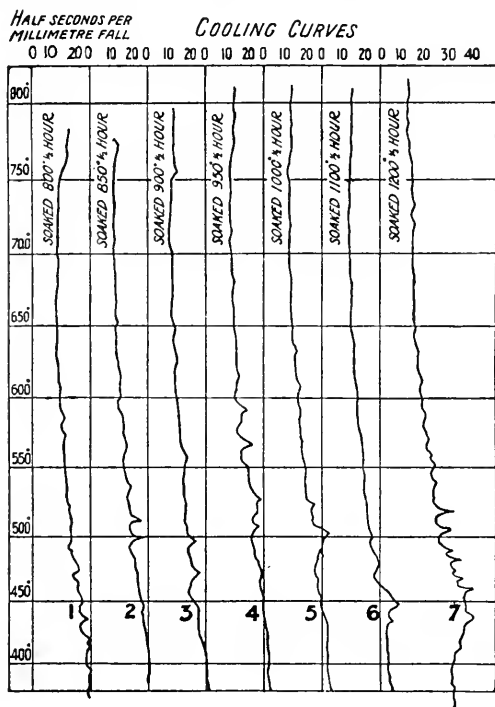


FIG. 15.—Steel No. 15. Carbon, 0.45 per Cent. ; Molybdenum, 8.11 per Cent.

temperature does not appreciably affect the position of the point. Curves 3, 4, 5, and 6 on Steel 6, and also Curves 13 and 14 on Steel 11, show that soaking just beyond the point will bring about lowering, where a quick heating to the same temperature does not give such a low point. The same is seen in Curves 4, 5, and 6 on Steel 3, although the temperatures in this case are not quite identical.

We see, therefore, that if the lowering temperature has not been exceeded, soaking has no influence on the critical points. If the lowering temperature has been exceeded, the tendency is for soaking to accentuate the lowering.

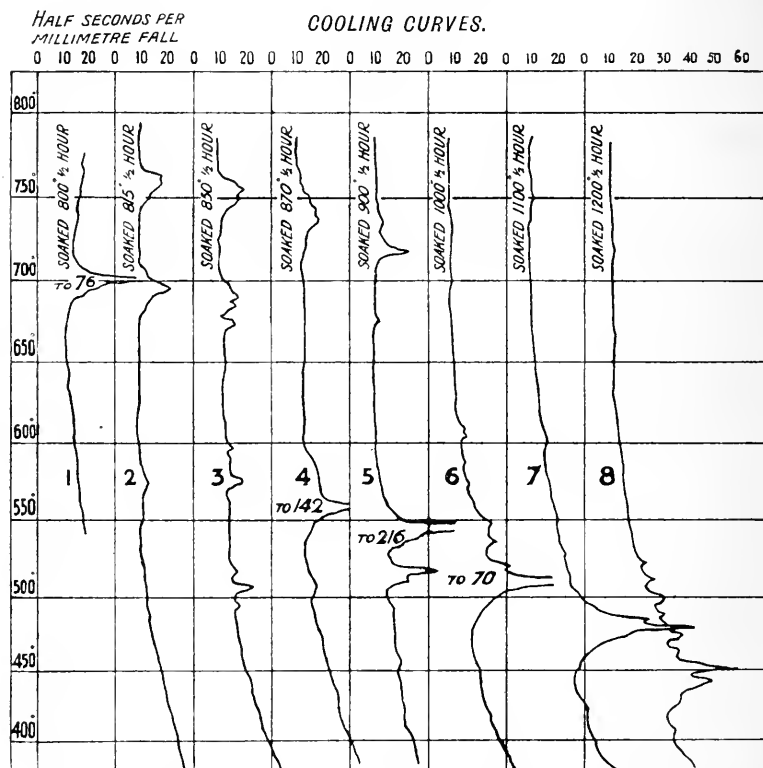


FIG. 16.—Steel No. 16. Carbon, 0·78 per Cent.; Molybdenum, 7·85 per Cent.

4. Influence of Rate of Cooling.

This is a subject which was dealt with very thoroughly in the tungsten series, previously referred to, with the result that rate of cooling had practically no influence on the position of the points.

Curve 24, Steel 3, was taken specially to test this point, and it will be seen that after heating to 1200° the piece was

cooled very slowly down to 800° , which temperature was maintained for two hours before cooling. The usual low point was found, showing the retarded cooling to have practically no influence on the position of the point.

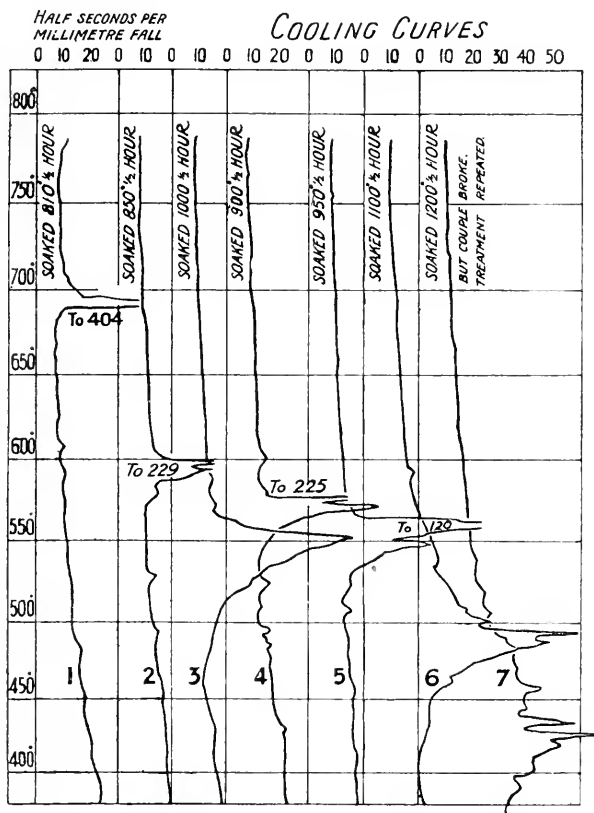


FIG. 17.—Steel No. 17. Carbon, 1.13 per Cent. ; Molybdenum, 7.92 per cent.

5. Influence of Carbon Content, Molybdenum being Constant.

In Fig. 19 are set out graphically the recalescence points and lowering temperatures for the different steels. Some difficulty is experienced in completing the diagram in the case of the high molybdenum steels, since no normal Ar1 has been found, and it would appear as though the lowering

temperature was as low, if not lower, than the Ac point. Thus Steels 13, 14, and 15 show signs of recalescence in the neighbourhood of 450° , and no point about 700° , even when cooled from 810° . This fact has been checked on duplicate and triplicate samples of each steel, and repeated heating to 800° , with one hour's soaking, does not produce a normal Ar1 point. It will also be recognised that the line MLT, denoting

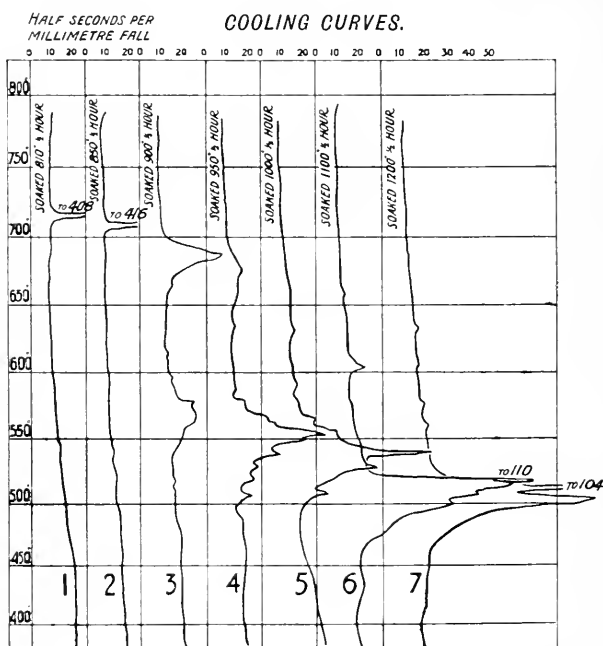


FIG. 18.—Steel No. 0. Carbon, 1.36 per Cent.; Molybdenum, 2.54 per Cent.

the first temperature which produced what was proved to be the maximum lowering effect, is somewhat arbitrary, and will depend upon the time of soaking.

Stated briefly, the chief points are:—

1. The normal Ar1 produced by cooling from immediately beyond the Ac point is practically constant about 700° to 710° C.

2. A point corresponding to Ar2 occurs in the 0.20 carbon series, and Ar3.2 in the 0.45 carbon series, and is practically

constant throughout the set at 760° . The former persists even after heating to 1200° (as in Steel 1), but the latter is usually dislodged after about 950° .

3. The position of the lowered point is absolutely independent of carbon content in any series.

4. As before noted, the lowering temperature—that is, the temperature first affecting the movement of Ar1—is difficult

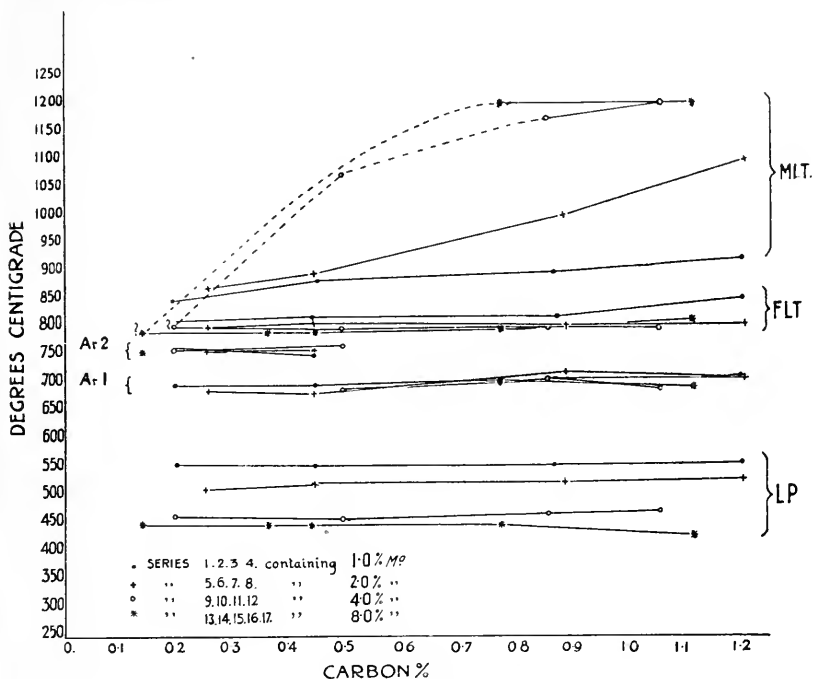


FIG. 19.—Diagram of Critical Points.

to fix definitely, especially when the molybdenum exceeds 2.0 per cent. In the first series it is clear that in Steel 4 the point is more difficult to lower than in Steel 3 or the still lower carbon members; but in the other series affected at such low temperatures that it can only be said that even with 1.12 per cent. carbon and 8.0 per cent. molybdenum the point is very decidedly lowered if the steel is heated to 850° .

5. The temperature which first produced the maximum lowering effect obtainable from the steel increases with the

carbon content, but this is much more marked in the presence of high molybdenum content. Thus in Steels 13, 14, 15 the low point first appears at what proves to be the lowest limit produced (by 1200° as maximum heating temperature); Nos. 16 and 17 give a lower point with each increment in initial temperature, and it is by no means certain in this series that 440° would be the limit if the heating were carried to beyond 1200° .

6. *Influence of Molybdenum Content, Carbon being Constant.*

1. The positions of the normal Ar1, Ar2, and Ar3.2 points are practically unaffected by molybdenum.

2. The position of the lowered recalcence depends on the molybdenum content, as shown in Fig. 20. The curve is

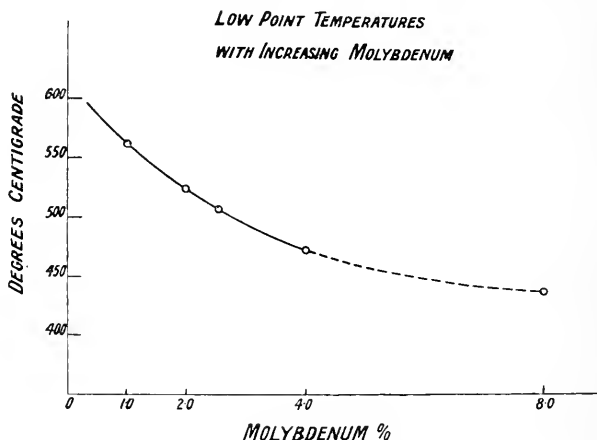


FIG. 20. Low Point Temperatures.

dotted beyond 4 per cent. molybdenum, since possibly, had the initial temperature exceeded 1200° , the low point would have been lower than 440° .

3. The "first lowering" temperature is slightly lower as the molybdenum increases.

4. The maximum lowering temperature is much higher as the molybdenum increases; thus Steel 4 is not affected until it has been heated to beyond 850° , but 1000° gives the

maximum lowering, whilst Steel 12 is affected by about 810° , and yet requires 1200° to produce a maximum effect.

(B) THE RECOVERY OF A NORMAL POINT AFTER LOWERING HAS TAKEN PLACE.

This depends very largely upon whether the point has been completely lowered or not. If, as in the case of Steel 3, Curves 2, 3, 4, 6; Steel 7, Curves 13, 14, 15, 16; and Steel 11, Curves 7, 8, 9, 10, the lowering temperature has only been slightly exceeded, then subsequent reheating to beneath this temperature will immediately restore a normal point.

If, on the other hand, the steel has been up to 1200° , and a large, low point obtained, the restoration is somewhat complicated.

Steels 3, 7, and 11 have specially been examined in this direction. Taking first Steel 3, Plate XII., it is seen that quick reheating with soaking nil (Curves 11, 17, and 19), or soaking half an hour (Curve 15), or soaking one hour (Curve 21), the point is slightly lifted, but no recalescence is found at the normal temperature. Again, with fairly slow reheating and soaking nil (Curves 13 and 20), or soaked one hour (Curve 22), the normal point is not recovered, although the low point is lifted slightly.

However, if after a quick reheating without soaking the piece is again reheated slowly without soaking, a normal point is recovered (Curves 11 and 12). Also, if after a slow reheating without soaking the piece is again reheated quickly without soaking, the point is broken up and partly recovered (Curves 13, 14). Similarly, a quick reheating with half an hour's soaking, followed by a quick reheating and no soaking, partly recovers the point (Curves 15, 16).

In order to test the matter further, a piece was reheated quickly without soaking three times successively (Curves 17, 18, and 19), and it is seen that the third reheating completely recovered the point. From this it would appear as though the reverse of the low point occurred, at any rate in part, during the reheating process, and a further pair of curves, 25 and 26, were taken for information on this par-

ticular point. The heating in each case was extremely slow, taking three hours to reach 800° ; No. 25 was soaked for a further two hours, and No. 26 was cooled immediately. The former has partly restored the normal point on a smaller scale, and the latter has broken up the low point, but has not restored a recalescence at the normal temperature.

A further trial was made by reheating very slowly up to 540° , keeping between 540° and 570° for one hour, and then proceeding smartly to 800° , and cooling without soaking. The resulting curve (23) still showed a fairly large point at the lowest limit. In this series Curves 1-16 were all down on one piece, Curves 17, 18, and 19 on another piece, and Curves 20-26 each had a new sample. Steel 7, Fig. 7, was similarly examined in Curves 7, 8, and 9, and showed that in this case rapid reheating with two hours' soaking gave a distinct but small point about 685° , whilst on a second slow reheating with half an hour's soaking, a large normal point was recovered.

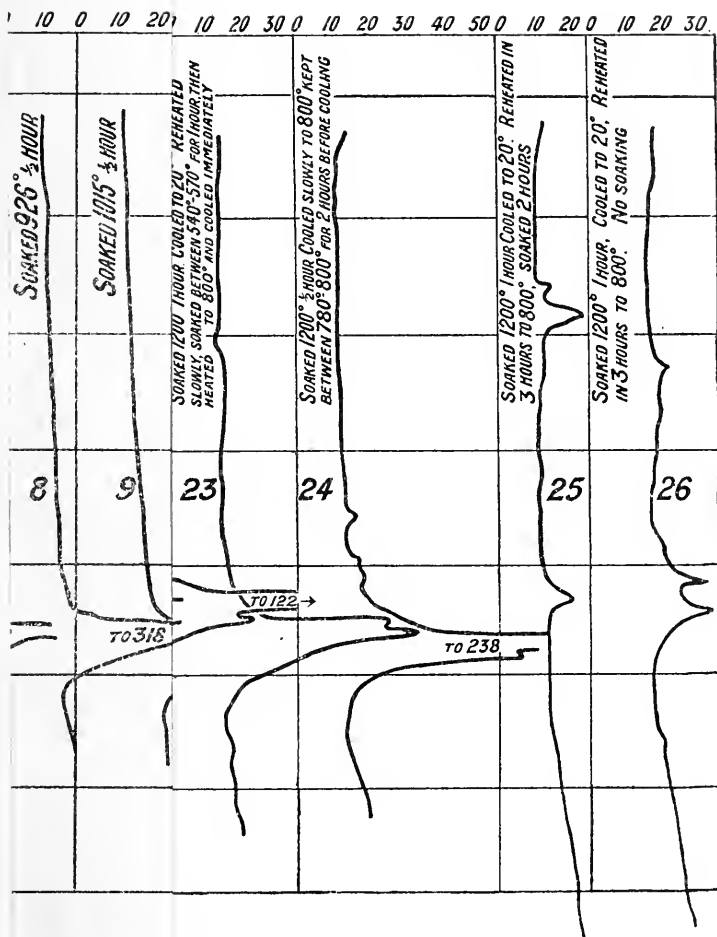
The chief feature of the recovery as shown by these series is the efficiency of repeated reheating rather than slow reheating or prolonged soaking.

(C) HEATING CURVES (Plate XIII.).

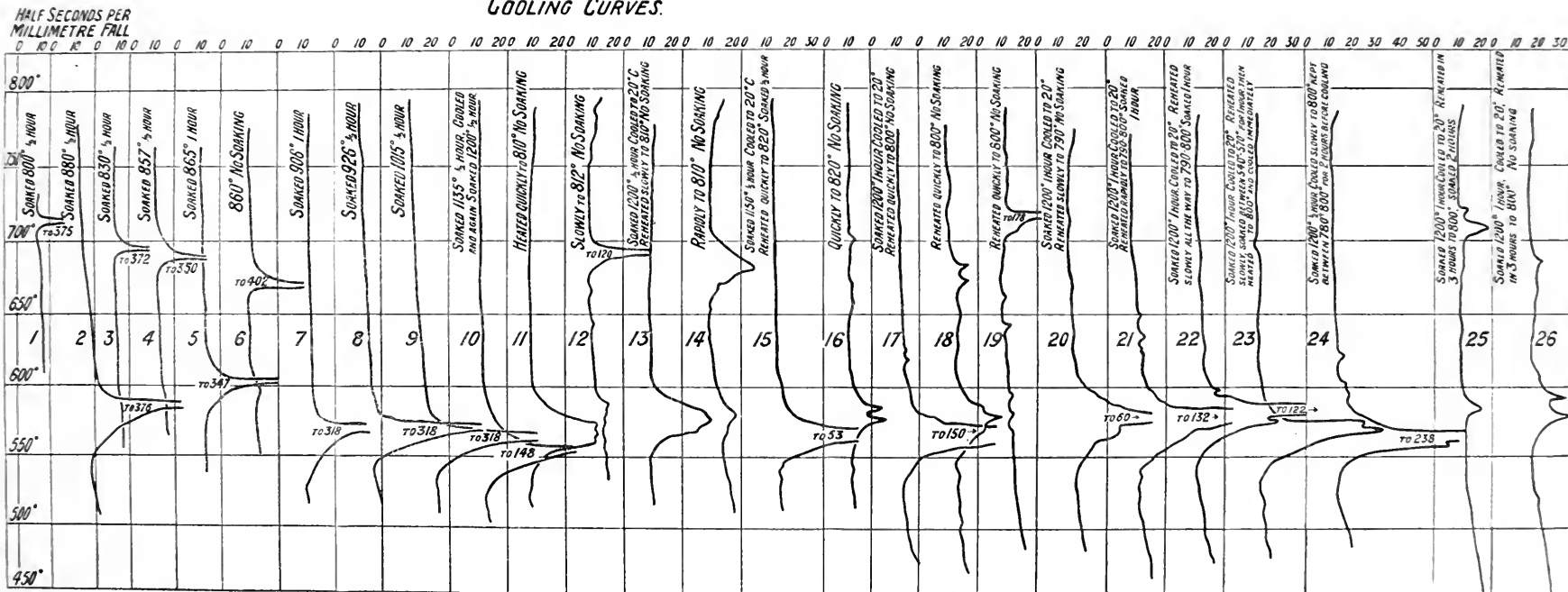
The above phenomenon led to an examination for possible recalescence points in the range of temperature below the ordinary Ac heating point. Steel 3, Heating Curve 22, Plate XIII., shows the heating curve from 250° upwards, after previously cooling from 1200° . A very small point about 610° is the only sign of a recalescence.

Heating curve 4B2 taken on Steel 4, after being cooled from 800° , shows no recalescence from 250° upwards until it reaches the normal Ac point.

A thorough examination has also been made of the heating curve above the Ac point in order to detect the presence of the recalescence point corresponding to the lowering temperature. Seeing that the "first lowering" temperature is in the neighbourhood of 800° , this was much more favourable for

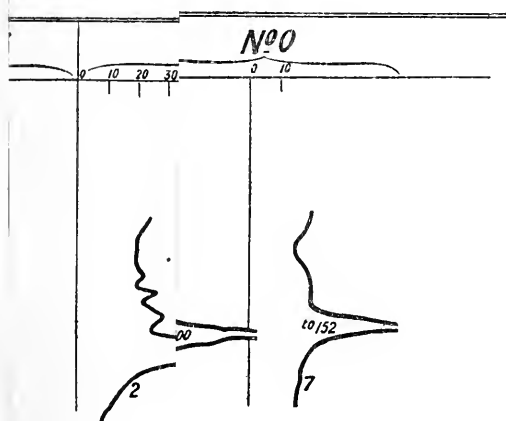
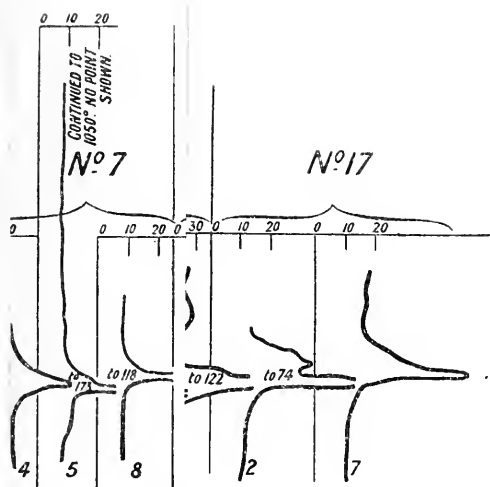


COOLING CURVES.



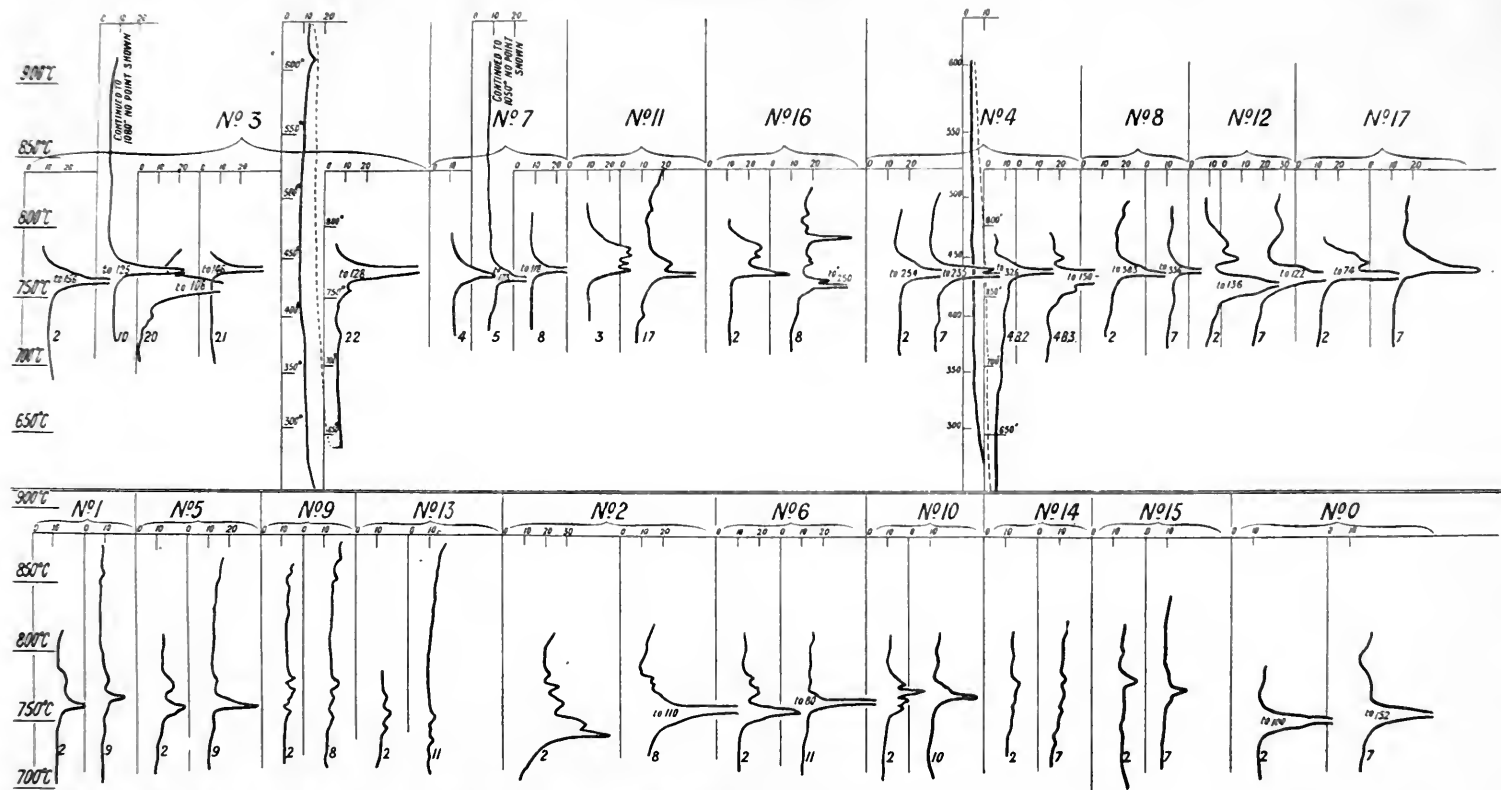
Steel No. 3.—Carbon, 0·87 per Cent.; Molybdenum, 1·02 per Cent.

PLATE XIII



G CURVES. THUS
FOR N°2 COOLIN
BTAINED FROM 7

HEATING CURVES.



THE HEATING CURVES NUMBERS CORRESPOND TO THOSE FOR COOLING CURVES. THUS "N°2" HEATING CURVE IS TAKEN AFTER COOLING CURVE N°1 & DURING REHEATING FOR N°2 COOLING. THE PREVIOUS INITIAL TEMPERATURE IS THEREFORE TO BE OBTAINED FROM THE PRECEDING NUMBER IN COOLING CURVES.

the purpose than were the tungsten steels previously examined, where the lowering temperature was nearer 1000° .

Steel 3, Heating Curve 10, taken after cooling from 1015° , showed no point above Ac1.2.3 up to 1080° . Similarly, Curve 5, Steel 7, showed no point above Ac1.2.3 up to 1050° . The latter had previously been cooled from 800° .

The heating curves are plotted in batches of similar carbon content, and each steel is at least represented by a pair of curves taken after previously cooling from below and above the lowering point. A noteworthy feature of the low carbon batch (1, 5, 9, 13) is the diminution in size of the Ac1 with increasing molybdenum, that is, Steels 1 and 9 of identical carbon content. Ac2 is fairly clear about $760-770^{\circ}$, but Ac3 has not been clearly distinguished. Ac1 occurs a few degrees higher than normal when the steel has previously shown the low point on cooling.

Similar characteristics are met with in the batch 2, 6, 10, 14, and 15 containing about 0.45 per cent. carbon. The point is extremely small in the high molybdenum steels, and gradually rises from 745° to 780° . Ac2.3 is seen clearly in Steels 2 and 6, but the point is small, and is associated with Ac1 in Steels 10, 14, and 15.

Five heating curves are given for Steel 3. Nos. 2, 20, and 21 represent normal, slow, and quick heating through the point. The point occurs about 10° higher, after a previous low-point cooling curve. Steels 7 and 11 do not show much distinction in the pair of curves, but No. 16 shows a curious smaller point which is new to this steel.

Steels 4, 8, 12, 17, and 0 with high carbon show a large single point which occurs a few degrees higher, after a previous low-point cooling curve.

SUMMARY OF OBSERVATIONS ON RECALESCENCE CURVES.

1. If the initial temperature has not exceeded a certain definite minimum the steel behaves in every way like plain carbon steel, except that the 0.20 and 0.45 carbon steels show very small points as the molybdenum increases. This "lowering temperature" is very low in molybdenum steels

compared with tungsten steels, and certain facts point to it being below the Ac1 point in the 8 per cent. series of low carbon.

2. If the "lowering temperature" has been exceeded, the normal Ar1 point is lowered or split, and as the temperature increases assumes a definite low position. The results show that there is a certain intermediate zone in some cases (as Steel 4) in which an equilibrium exists between the two points even with fairly slow heating and two hours' soaking.

The position of the lowered recalescence is controlled by the molybdenum content. The critical point is more easily affected as the molybdenum content increases, but a higher temperature and longer soaking is necessary to produce the full lowering effect. The temperature which first affects the point, and also that which produces the maximum lowering effect, is slightly higher as the carbon content increases.

The normal point is not recovered by soaking during cooling, at a temperature which ordinarily gives the normal point, if the steel has previously been heated to 1200°.

The recovery of the normal point after the low point has been produced is of great interest, and there are yet certain aspects of the matter which require further elucidation. From the previous work on tungsten steels, and from the experiment just quoted, it is clear that the normal Ar1 cannot be recovered until the low point has been passed in cooling. The quickest way of effecting a complete recovery is to reheat two or three times to a temperature below that producing lowering. Even quick heating without soaking will recover the point if repeated several times.

Referring to special experiments on Steel 3, we find, on the other hand, that if the first reheating after lowering is extremely slow and the piece is then cooled without any soaking, the normal point is by no means fully recovered. Even soaking at 540° to 570° for an hour during reheating does not recover a normal point. Extremely slow heating, accompanied by two hours' soaking, does effect a partial recovery.

The Steel 7 gave signs of easier recovery than 3, but the same principles are supported.

This is somewhat contrary to the author's experience with tungsten steels, where it was found that a single reheating to below the lowering temperature very largely recovered the normal point, but generally left traces of the low point.

Heating.—In curves taken on Steels 3 and 4 from 250° upwards no distinct recalescence has been noted beyond the ordinary Ac point either if cooled previously from a high or low temperature. Neither has a point been noted above the Ac, up to 1050°, although in some cases there is a broadening out of the upper limit of Ac1 which is actually near the lowering temperature. However, a careful general study of the heating curves does not confirm this being associated definitely with the lowering point. The most remarkable feature of the heating curves is the diminution in size of the Ac1 point as molybdenum increases. The position of Ac1 is not greatly affected by the previous treatment, but is slightly higher when previously cooled from above the lowering temperature.

The possible meaning of the phenomena observed in the heating and cooling curves will be discussed later, together with the resistivity, microscopic, and hardness data.

ELECTRICAL RESISTIVITY.

In the author's first report (*loc. cit.*), p. 102 *et seq.*, are given values for the electrical resistivity of this series of steels in five different conditions—drastically annealed, lightly annealed, normalised, as rolled, hardened and tempered. The determinations were made upon a Thomson double bridge of extreme accuracy, described on p. 103.

The resistivity is of very considerable interest in investigating the constitution of the steels, and these results will be referred to in detail. A summary of these results will be found in Table II., together with further values for quenched steels to be described later, and theoretical values are included, calculated according to Benedicks' formula previously discussed (p. 102).

Benedicks showed that the resistivity was increased propor-

TABLE II.—*Electrical Resistivity.*

Mark.	Chemical Analysis.		Carbon Equivalent of Impurities.						Calculated Resistivity by Benedicks' Formula.						Actual Resistivity, Microhms per Cubic Centimetre.						
	C per Cent.	Mn per Cent.	C mill.	C = q.s. ¹	C mill.	Mn in Solu- tion.	C = q.s. ¹	Mn in Solu- tion.	All C.	All Mo in Solu- tion.	All C.	All Mo in Solu- tion.	All C.	All Mo in Solu- tion.	"Carbide"	Annealed.	Normalised.	Rolled.	Hardened and Tempered. ²	Quenched 800°.	Quenched 1200°.
			1	2	3	4	5	6	1	2	3	4	5	6	1	2	3	4	5	6	7
1	0.195	1.030	0.084	0.194	0.213	0.373	0.279	0.408	9.8	12.8	13.3	16.2	15.0	18.6	...	13.48	14.51	14.18	14.22	16.0	...
2	0.445	1.054	0.102	0.262	0.234	0.384	0.547	0.679	10.3	14.6	13.8	17.9	22.3	25.8	...	14.77	16.55	17.40	16.86	21.0	...
3	0.869	1.018	0.128	0.398	0.255	0.525	0.997	1.124	11.0	18.3	14.4	21.7	34.3	37.7	17.46	18.55	20.47	21.00	20.25	36.3	...
4	1.215	1.006	0.122	0.392	0.259	0.529	1.337	1.474	10.8	18.1	14.5	21.8	43.4	47.1	18.65	19.54	20.32	23.10	21.15	39.53	45.31
5	0.246	2.176	0.090	0.210	0.362	0.482	0.336	0.608	10.0	13.2	17.3	20.7	16.6	23.9	...	13.55	16.73	18.66	17.93	16.5	18.96
6	0.442	2.181	0.105	0.255	0.378	0.528	0.547	0.820	10.4	14.4	17.8	21.8	22.2	29.5	...	14.97	16.42	20.66	19.47	20.1	24.82
7	0.883	2.186	0.107	0.377	0.380	0.650	0.990	1.263	10.4	17.7	17.8	25.0	34.1	41.4	15.85	16.13	18.80	22.90	20.10	33.44	...
8	1.210	2.109	0.099	0.369	0.363	0.653	1.309	1.573	10.2	17.5	17.3	23.1	42.7	49.8	17.76	18.47	21.72	24.80	20.80	38.52	48.66
9	0.190	4.110	0.082	0.192	0.596	0.706	0.272	0.786	9.7	12.7	23.6	26.5	14.9	28.7	...	14.95	16.40	20.22	19.12	16.82	23.66
10	0.487	4.009	0.094	0.254	0.595	0.755	0.581	1.082	10.1	14.4	23.6	27.9	23.1	36.5	...	14.89	16.39	24.60	19.52	19.54	34.04
11	0.865	4.002	0.087	0.357	0.587	0.856	0.952	1.452	9.9	17.2	23.3	30.5	33.1	46.5	15.15	15.53	18.42	25.70	19.88	26.95	36.42
12	1.060	4.019	0.083	0.353	0.585	0.855	1.143	1.645	9.8	17.1	23.3	30.5	38.2	51.7	16.18	16.92	19.37	27.10	19.78	36.03	...
13	0.135	8.012	0.084	0.179	1.085	1.180	0.219	1.220	9.8	12.4	36.6	39.1	13.5	40.2	23.48	23.70	25.76	25.83	25.25	25.23	32.73
14	0.361	8.167	0.099	0.244	1.120	1.365	0.460	1.481	10.2	14.1	37.6	44.2	19.9	47.3	...	15.32	17.20	19.00	18.79	17.2	34.03
15	0.445	8.109	0.106	0.376	1.120	1.370	0.551	1.565	10.4	14.5	37.6	44.3	22.4	49.6	...	14.38	16.79	28.00	19.06	17.5	37.38
16	0.775	7.857	0.083	0.353	1.064	1.384	0.858	1.839	9.8	17.1	36.1	43.3	30.6	56.8	15.28	15.38	17.87	27.40	19.92	22.07	46.07
17	1.125	7.920	0.100	0.370	1.090	1.360	1.325	2.215	10.2	17.5	36.8	44.1	43.1	66.5	16.85	16.13	18.86	22.60	19.50	36.2	55.57
0	1.360	2.540	0.115	0.385	0.432	0.703	1.475	1.792	10.7	17.9	19.2	26.4	47.1	55.7	...	20.13	21.79	31.30	22.40	41.5	...

¹ "q.s." represents the quantities *satisfaisantes* used by Benedicks.² Hardened in oil from 650° for 0.2 C, 900° for 0.45 C, 850° for 0.9 C, 800° for 1.2 C, and tempered together in lead at 550° for 15 minutes.

tionally to the amount of dissolved impurity in iron, and that equiatomic solutions gave the same resistivity. The formula used is—

$$\sigma = 7.6 + 26.8 \Sigma C \text{ where } \Sigma C \text{ represents the sum of the carbon equivalents,}$$

i.e. carbon + $\frac{1}{2}\%$ silicon + $\frac{1}{8}\%$ manganese + &c.

In discussion of these results it was pointed out (p. 111) that in the case of annealed steels, molybdenum did not increase the resistivity, in fact in some cases there was a slight reduction. It was therefore stated that the molybdenum did not exist in the state of solid solution in the iron. Attention was drawn to the high value for No. 13, which at that time appeared inexplicable.

Closer examination of the results reveals some extremely interesting facts which are illustrated in Fig. 21.

Considering the steels in the annealed condition, in series with carbon, manganese, and other elements constant, and with increasing molybdenum in steps of 1, 2, 4 to 8 per cent., it is seen that the resistance—

1. Is increased in the case of the mildest steels (1, 5, 9, 13) as the molybdenum content is increased.
2. Is practically constant in the second series with about 0.45 per cent. carbon (Steels 2, 6, 10, and 15).
3. Is generally decreased in the third series, with about 0.9 per cent. carbon (3, 7, 11, 16—the latter is somewhat low, with 0.78 per cent.).
4. Is decreased in the fourth series. These higher carbon series (4, 8, 12, 17) are not quite regular in carbon, but making allowances for this the resistivity certainly falls as the molybdenum content increases.

These results are the mean of closely agreeing determinations on three separate bars, heated to about 950° C. for four hours and cooled over about sixty hours.

A series of tests of the higher carbon steels after a more drastic annealing (for carbide analysis) by keeping for ten hours at about 950° C., and cooling over about fifty-five hours, showed a slight reduction in the resistivity all round, more particularly in the lower molybdenum steels. Even then the same statements hold. Note, for example, Steels 3 and 11 with almost identical analysis except for molybdenum, and the specific

resistances are respectively 17.46 and 15.15 microhms; also the high value for No. 13 is maintained.

Bars normalised from 900° C. show similar characteristics, but the values are all slightly higher than the annealed bars.

Having completed the cooling curve data furnishing the critical change points, it was decided to examine the resistivity of samples quenched—

1. From below the "lowering temperature."
2. From above the "lowering temperature."

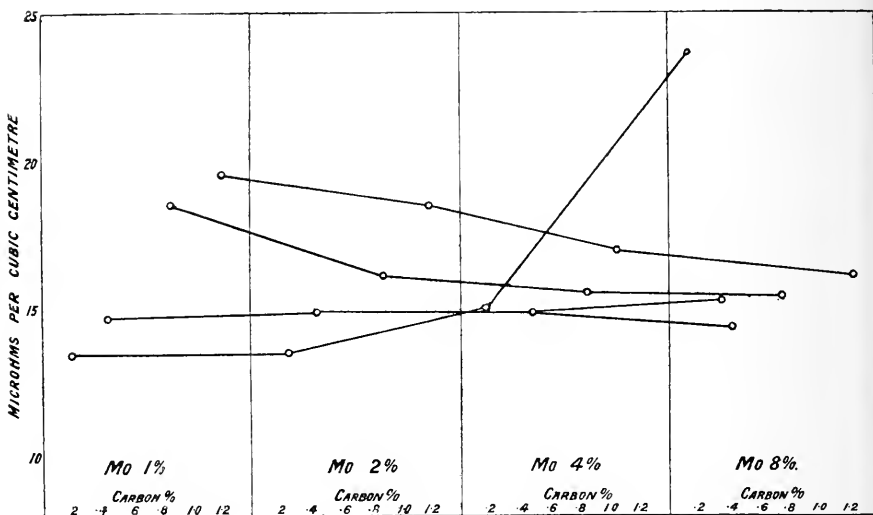


FIG. 21.—Electrical Resistivity of Annealed Steels.

Bars 4 inches long, $\frac{5}{8}$ inch diameter, were first normalised from 900° C. and then turned down very accurately to $\frac{1}{8}$ inch diameter and polished. The small section was chosen so that quenching should be as effective as possible throughout the specimen. The bars were heated up fairly slowly and soaked thirty minutes before quenching.

The heating was carefully performed in the furnace used for the recalescence work, every precaution being taken to heat the bars evenly, and the same couple used as for the recalescence curves. A current of pure dry nitrogen was passed over the samples whilst in the furnace, and there was no trace

of scale after quenching. The quenching liquid was water at about 20° C. After quenching the bars were lightly polished with "Fortin" 0 emery cloth, and the resistivities determined on 5 centimetre length as elsewhere described.

The last two columns of Table II. give the results obtained.

Quenched at 800°.

These values form a very interesting series. A comparison of these figures with those given in column 5 for the calculated resistivity, allowing all the carbon and no molybdenum in solution, shows that in Series 1-4 and 5-8 there is almost perfect agreement, allowing for a little free cementite in 4 and 8. The increase in resistivity in Steels 10, 11, and 12 is not as great as steels of lower molybdenum and equal carbon. Steel 13 is only slightly above the annealed value, and the other members of the 8 per cent. series show only small increases, and except for No. 13 actually do not reach the calculated value for all carbon and no molybdenum in solution. As will be seen from the microscopic data presented later, Steels 11, 12, 16, and 17 all show specks of free cementite after quenching from 800°.

The conclusion to be drawn from these figures is that below the "lowering temperature," the carbon is in solution almost completely in the 1 and 2 per cent. molybdenum steels, and less completely as the molybdenum increases. The increase in resistivity by this quenching is less (for the same carbon content) as molybdenum increases, and there is nothing to favour the view that molybdenum has entered into solution.

A further point of interest is found in the figures for steels hardened in oil from 950°, 900°, 850°, or 800° (according to carbon content) and tempered for ten minutes in lead at 550°. The resistivity is brought back by this tempering almost to the normal state.

Quenched at 1200°.

Some difficulty arises here owing to the bars cracking in quenching, and several of the bars were lost on this account. It is sufficient to note, however, that in the cases where good determinations have been made, there is a substantial increase

in the resistivity, and also that the increase is greater (where carbon content is similar) as the molybdenum content increases. Further, in every case, and clearly in the 8 per cent. molybdenum series, the resistivity is greater than can possibly

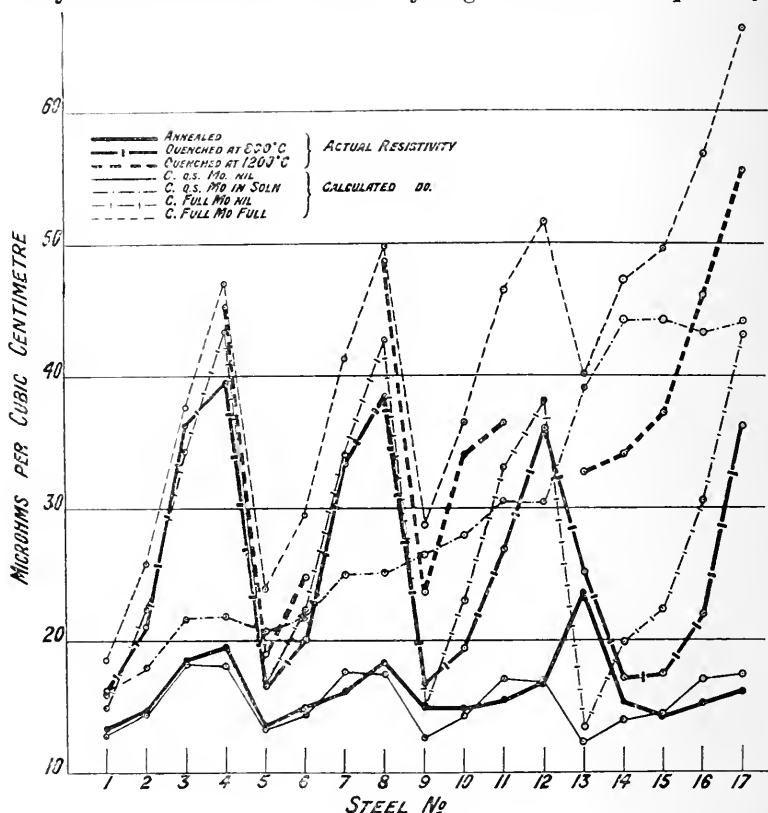


FIG. 22.—Electrical Resistivity.

be accounted for by all the carbon and no molybdenum in solution.

The author would suggest that in this state the steels contain at least a portion of the molybdenum in the state of solid solution.

The full series of determinations, together with some theoretical values, are plotted in Fig. 22, and clearly illustrate the points noted.

MICROSTRUCTURE OF QUENCHED SPECIMENS.

In the author's earlier report on this subject, rolled, annealed, normalised, hardened (oil 950° – 800°) and tempered (550°) bars were described. The chief facts noted were—

1. The pearlite is very emulsified even in the annealed steels.
2. The "saturation point" beyond which excess carbide appears is lowered as the molybdenum increases.
3. No special constituent is present, but the excess carbide assumes a peculiar worm-like appearance as illustrated in Steel 4 annealed (Plate XV.).
4. Normalising from 950° shows the cementite to be partially soluble in the martensite, but not very readily.

The specimens used for quenching experiments were $\frac{1}{4}$ inch diameter by $\frac{1}{8}$ inch thick, and were treated in the furnace used for the recalescence work. Nitrogen was passed through the tube during the heating operations, and there was no scaling. The samples were quenched singly by withdrawing in a "combustion" boat, and tipping into a large water-bath at 16° C. The surface was ground off on a wet stone to a depth of $\frac{1}{32}$ inch, and the specimen then polished and etched.

Endeavour has been made to prepare the specimens in such a way as to reveal the molybdenum selectively by the action of potassium sulpho-cyanide in reducing solution; silver nitrate in nitric acid; sulphuretted hydrogen in hydrochloric solution; and in other ways; but the results are not sufficiently conclusive to make their inclusion advisable in this paper. The structures to be described have been obtained by thirty seconds' etching in 5 per cent. picric acid in alcohol, and two pairs of steels (2 and 16, 4 and 17) have been photographed in each condition to show typical structures (Plates XIV. to XVII.).

Quenched at 800° .

Steels 1 and 2 show pale yellow martensite areas in a ground of ferrite, No. 3 dense martensite, and No. 4 martensite with a few fine specks of cementite.

Steels 5 and 6 resemble 1 and 2. No. 7 shows martensite with fine specks of cementite, whilst No. 8 contains an appreciable quantity of fine cementite in cellular formation.

In Steels 9 and 10 the ground mass etches up light brown with small rounded areas of lighter tint. No. 11 shows specks of cementite set in pale martensite, and No. 12 is similar but with more cementite.

No. 13 shows very pale martensite which is very easily pitted; Nos. 14 and 15 dark, fine grey martensite. Nos. 16 and 17 show numerous specks of cementite in a dark matrix of martensite.

The structures are therefore of a similar character to those met with in carbon steels, consisting of martensite in which the free carbide has partially but not wholly dissolved in the mass.

Quenched at 1200°.

All the steels from 1 to 10 show pale, even martensite figured lightly with fine brown etching lines. The only noticeable feature between the different steels is that the higher carbon steels etch just a little darker.

Steels 11 and 12 show a pale, yellow, polyhedric structure with annular brown etching marks. Steel 13 shows a white ground with numerous rounded areas, presumably of martensite, which etch up light brown. No. 14 is practically structureless, whilst 15 shows a fairly dark martensite ground, with small irregular patches of lighter tint. Steel 16 shows pale polyhedra with specks of cementite, and 17 is similar but with the rounded cementite specks in greater profusion. It is interesting to note that these are the only two specimens showing free cementite in this series.

Heated to 1200°, Cooled down to 600°, and then Quenched.

The 0.45 and 0.9 carbon members of the series have been examined in this way and are of great interest.

Heated to 1200° and slowly cooled down to 600°, *i.e.* below the normal Ar₁ but above the low point, all the sections correspond to hardened steels. Etched along with the specimens previously described, Nos. 2, 6, 3, and 7 all showed

very pale martensite with etching marks. Nos. 10 and 11 have developed white polyhedra, the latter with extremely fine lines. Nos. 15 and 16 similarly show polyhedra, the latter with fine specks of free carbide. Plates XIV. to XVII. show Nos. 2, 3, 15, and 16 of this series.

AIR-COOLED SPECIMENS.

A selection of specimens of the same size as the quenched ones were heated to 1200° for half an hour, and were then withdrawn and allowed to cool naturally on a piece of gauze.

Steel No. 1 shows a close matrix of ferrite and pearlite. The Steels 2 and 4 show distinct signs of self-hardening. The structure strongly resembles those published by the author¹ for air-cooled tungsten steels, No. 2 showing drab troostite in angular masses enclosing pale martensite, whilst No. 4 shows less troostite, more martensite, and very little free cementite.

No. 13 shows rounded troostite areas in ferrite, No. 14 dense troostite martensite, No. 15 a pale ground of martensite with troostite in fine black needles and numerous fine white specks, probably cementite. Nos. 16 and 17 are similar to 15, except that the excess cementite appears more and more abundantly.

Photomicrographs have been prepared and are reproduced in Plates XIV. to XVII., showing Steels 2, 4, 15, and 17 in the annealed, air-cooled from 950° , quenched 800° , quenched 1200° , air-cooled from 1200° , also Steels 2, 3, 5, 15, and 16 after heating to 1200° and quenching at 600° . The micrographs are reproduced as typical to illustrate the description given in the paper. The magnification is uniformly 150 diameters.

The most important points revealed by the heat-treated specimens are as follows:—

1. Specimens quenched from 800° show normal martensite and show that the carbide is partly soluble at this temperature.
2. Quenching from 1200° the steels are less easily attacked by acid and, in the case of certain steels

¹ *Journal of the Iron and Steel Institute*, 1909, No. 1., Plates XV., XVI., XVII.

with the higher carbon and molybdenum contents, show white polyhedra. Free carbide is shown in Steels 16 and 17 after this treatment.

3. Heated to 1200° and quenched at 600° , the structures are essentially the same as those quenched at 1200° .
4. The samples air-cooled from 1200° show martensite and troostite, the proportion of martensite being greater as the carbon and molybdenum contents increase.

HARDNESS TESTS OF QUENCHED SPECIMENS.

In the previous report the author showed that with steels in the annealed state, crushing, Brinell, and scleroscope hardness tests, supported also by tensile and bending tests, showed there was no increase in hardness as the molybdenum content increased.

Hardness tests have been made by the Brinell method, using 3000 kilogrammes load on a 10-millimetre ball, on samples quenched from 800° and 1200° , and air-cooled from 1200° . The harder samples quenched from 1200° were brittle and cracked under the load, so that reliable figures cannot be included; also it will be understood that values beyond about 580 are probably affected somewhat by the flattening of the ball. However, the results presented in Table III. are of considerable interest. They show very clearly that, in the series quenched from 800° , higher molybdenum does not increase the hardness, carbon being constant. Note; for example, Steels 2, 5, 10, 15, all containing about 0.45 per cent. carbon, but increasing contents of molybdenum. Samples quenched from 1200° all show a considerably increased hardness, and the increase is particularly noticeable in the 8 per cent. molybdenum series. With constant carbon the tendency is for the hardness to increase with increasing molybdenum. This, of course, confirms the previous data to the effect that molybdenum plays a very different rôle in the steel below and above the "lowering temperature" discussed in the chapter on recalescence.

The air-cooled samples confirm the microscopic examination,

and show that the steels are appreciably self-hardening, and that the phase existing at 1200° is comparatively stable during air-cooling.

TABLE III.—*Brinell Hardness Tests on Quenched Specimens.*

Steel No.	Carbon per Cent.	Molybdenum per Cent.	Quenched 800° .	Quenched 1200° .	Air-Cooled from 1200° .
1	0.19	1.03	202	332	143
2	0.44	1.05	387	444	330
3	0.87	1.02	600	460
4	1.21	1.09	652	477
5	0.25	2.18	217	375
6	0.44	2.18	346	444
7	0.88	2.19	600
8	1.21	2.11	652
9	0.19	4.11	293	375
10	0.49	4.01	351	418
11	0.86	4.0	627
12	1.06	4.02	600
13	0.13	8.01	156	220	137
14	0.36	8.17	179	460	411
15	0.44	8.11	228	512	430
16	0.77	7.85	477	600	578
17	1.12	7.92	555
0	1.36	2.54	627	477

GENERAL DISCUSSION OF RESULTS AND THEIR BEARING ON THE CONSTITUTION OF MOLYBDENUM STEELS.

The literature on the subject of the constitution of molybdenum steels is very meagre and is by no means consistent. It is generally accepted that the action of tungsten and molybdenum on the critical points, electrical resistivity, and other properties are remarkably analogous, and it will therefore not be considered irrelevant to refer to previous work on tungsten steels for confirmatory data.

It is proposed to discuss the question in two stages:—

1. The condition of molybdenum in annealed steels.
2. The nature of the change at the lowering temperature.

1. THE CONDITION OF MOLYBDENUM IN ANNEALED STEELS.

Three hypotheses will be considered :—

- (a) The molybdenum exists as a double carbide of iron and molybdenum.
- (b) The molybdenum is in solid solution in iron.
- (c) The molybdenum exists uncombined with the carbon, but probably as an iron molybdenum compound not in solid solution in the iron.

(a) *Double Carbide.*

Guillet¹ described his researches on two series of molybdenum steels, and divided the steels into pearlite and double carbide. The separation is a straight line joining 2·5 on the molybdenum axis with 1·6 on the carbon axis of a constitution diagram. This is arrived at by microscopic examination of unhardened specimens, and mechanical tests are stated to confirm. This diagram was also presented later,² and Portevin,³ working on the same steels, states that the double carbide is found in steels containing above 5 per cent. molybdenum with 0·2 per cent. carbon, and above 2 per cent. molybdenum with 0·8 per cent. carbon.

There is no definite proof that the constituent referred to by these writers is really a double carbide, and many facts in other investigations clash with this view. In any case we may assume that below the molybdenum contents specified the double carbide does not exist, and yet there is nothing in recalcence or resistivity data to support such a demarcation in the condition of the molybdenum. Moreover, a series of very careful carbide determinations reported by the author (*loc. cit.*) failed to find any evidence of double carbide. The fact that the critical points in cooling curves from below the lowering point are identical in position with plain carbon steels (and are actually smaller as the molybdenum increases), in steels which would be classed as containing double carbide, also throws some doubt on the correctness of this view.

¹ *Le Génie Civil*, August 13, 1904.

² *Journal of the Iron and Steel Institute*, 1906, No. II. p. 12.

³ *Carnegie Scholarship Memoirs*, vol. i., 1909, pp. 276, 331.

Again, as previously stated in this Report, the annealed steels do not show increased hardness with increased molybdenum, nor do specimens quenched at 800°. This is opposed to the view that a hard double carbide is present as claimed by Guillet. The fact that the "saturation" point in microscopic examination is lowered does not necessarily prove the existence of a double carbide.

(b) *Molybdenum in Solid Solution.*

The resistivity values in the present paper prove conclusively that molybdenum does not exist in a state of solid solution in the annealed steels.

Le Chatelier,¹ is of the opinion, based on resistivity determinations, that tungsten and molybdenum are isolated in the mass of metal in the form of definite compounds.

Boudouard² also states that in the case of tungsten steels the state of the metal, whether hard, annealed, or normal, appeared to have no influence on the resistivity. This is extremely interesting, and is supported by the resistivity tests on the molybdenum steel bars quenched from 800° now reported upon.

This is an important fact, since the theory of the lowered recalescence hangs very largely on this; and it has been suggested by Mr. Edwards³ that in the case of high speed steels the tungsten is actually in solid solution in iron below the change point in heating, the carbon existing as carbide of iron.

The matter will be referred to again in discussing the lowering of the critical points, but the resistivity values appear to prove conclusively that this theory is untenable.

(c) *Molybdenum not in Solution and not as Carbide.*

The resistivity values favour the view that molybdenum is not in solid solution in the annealed steels. Carbide analyses go to prove that the carbon exists as carbide of iron (as suggested by Edwards (*loc. cit.*) and independently by the

¹ *Contribution de l'Etude des Alliages*, p. 418.

² *Comptes Rendus*, vol. cl., No. 111. pp. 1475-8.

³ *Journal of the Iron and Steel Institute*, 1908, No. 11. p. 130.

author¹ for tungsten steels), but no definite iron-molybdenum compound was isolated.

In discussing the constitution of tungsten steels, the author suggested as a tentative hypothesis, that the tungsten existed in the form of a tungsten-iron compound, probably Fe_3W , not in solid solution in the iron. The objection to this hypothesis, as pointed out by Professor Carpenter in correspondence, is that no evidence of this tungsten compound has been distinguished in microscopic examination. This is, of course, a serious objection, and needs careful consideration. In the case of the molybdenum steels, no special iron-molybdenum compound has been detected "isolated in the mass." However, cases are known where a constituent may appear homogeneous under ordinary microscopic examination and yet consist of ultra-microscopic particles not in simple solution, with the result that the electrical resistivity of the "solvent" is not increased. Such an example is found in troostite as described by Benedicks, which has the same electrical resistivity as pearlite, and yet behaves as a colloidal solution of Fe_3C in iron.

This will be referred to later in discussing the lowering of the critical points.

2. THE NATURE OF THE CHANGE AT THE LOWERING TEMPERATURE.

Dealing first with a brief résumé of previous work on tungsten steels, several theories as to the nature of this change have been formulated.

Bohler³ suggested that both the normal and the lowered point corresponded to the formation of "pearlite," but that after heating beyond about 1100° the tungsten suspended the formation until the low point temperature.

Osmond⁴ suggested that the duplication was due to segregation of cementite caused by the presence of tungsten (and chromium).

¹ *Journal of the Iron and Steel Institute*, 1907, No. II. p. 320.

² *Ibid.*, 1908, No. IV. p. 221.

³ *Wolfram und Rapid Stahl*, 1903.

⁴ *Revue de Métallurgie*, 1904, p. 348 et seq.

Carpenter,¹ dealing with high-speed steels, and a few tungsten and molybdenum steels, suggested that the action was probably due to the widening, splitting, or lowering of the normal critical range, and that mild quenching in an air-blast caused the complete or practically complete suppression of the critical range.

The author (*loc. cit.* 1907, No. I. p. 291) showed that both the lowering temperature and the low point temperature were of a definite character, and depended on the tungsten content. No recalescence was found at the lowering temperature.

Mr. Edwards² suggested that the tungsten, existing in solid solution in iron, slowly forms a carbide of tungsten at 1200° (for high speed steel), which carbide has a lower recalescence than has carbide of iron on cooling. Beyond this, in high speed steels a double carbide of tungsten and chromium is formed, which is not decomposed by cooling in an air-blast. The carbon exists as carbide of iron if the steel is not heated beyond the lowering temperature, and enters solution in the ordinary way about 770°. The lowered recalescence is not the carbide of iron change suppressed or lowered by tungsten, but is an entirely new point due to carbide of tungsten.

Professor Carpenter,³ speaking of high speed steels, said, "It seems probable that the unhardened tool contains carbide of iron just like the carbon tool." Referring to the completely new change between 540° and 460° when cooled from 1100° he said, "It is probable that the latter corresponds to the separation of a special carbide containing iron and either tungsten or manganese, or both, which gives the special self-hardening property to the self-hard tool."

The author pointed out (*loc. cit.* 1909, No. II. p. 245) what appeared to be an essential difference in these two theories, viz. that carbide of tungsten was not the same as a double carbide, but in correspondence (p. 253) Professor Carpenter and Mr. Edwards did not consider this an important difference, but "simply a question whether carbide of tungsten separated with or without iron or manganese." The main point is the same in both cases, therefore, that a carbide of tungsten is

¹ *Journal of the Iron and Steel Institute*, 1904, No. I. p. 433.

² *Ibid.*, 1908, No. II. p. 130.

³ Manchester Association of Engineers, November 28, 1908.

formed beyond the lowering temperature, and that below this temperature the carbon exists as carbide of iron, and in Edwards' theory, tungsten is in solid solution in the iron.

The author¹ dealt with these theories in the light of exhaustive research on 3 per cent. tungsten steels, and suggested that Edwards' theory was not tenable. A tentative hypothesis was formulated wherein the lowering of the recalescence point was due to the solution of a tungsten-iron compound (probably Fe_3W) at the lowering temperature. The tungstide is reprecipitated at the low point and Fe_3C immediately follows. In correspondence on this paper Professor Carpenter and Mr. Edwards drew attention to an admittedly weak point in the hypothesis, viz. that no special constituent corresponding to the separated tungstide of iron had been found on microscopic examination. They suggested that the presence of a tungstide should make the steel harder in the annealed state, whereas this is not the case; also that the separation of the tungstide at the low temperature should produce a recalescence independent of the presence of carbon.

The summary of work on tungsten steels therefore illustrates the essential differences in the theories published, and forms a very useful guide in discussing the constitution of molybdenum steels.

The author has given in the text of this paper the chief conclusions arrived at in the earlier report on the molybdenum steels now under investigation. The matter was then left open, but the majority of the evidence was in favour of the view that, as with tungsten steels, molybdenum exists in the annealed steel in the form of an iron-molybdenum compound, not in solution in the iron.

Turning to an examination of the data obtained in the report now presented, the author would submit that this on the whole is strongly in favour of an hypothesis of the constitution of carbon-molybdenum steels similar to that suggested for tungsten steels, with slight modifications. It is true that there are certain objections, but it is very probable that by further investigation these will be satisfactorily explained.

¹ *Journal of the Iron and Steel Institute*, 1909, No. II. p. 252.

The hypothesis is that molybdenum exists in the annealed steels in the form of an iron-molybdenum compound, not in simple solution in the iron. On heating the Fe_3C goes into solution at $\text{Ac}1$, and, unless the "lowering temperature" is exceeded, is reprecipitated at the normal $\text{Ar}1$ on cooling. If heated to beyond the lowering temperature the condition of the molybdenum compound undergoes a change, and behaves in such a way as to suggest that it has entered into solution in the iron. The separation of Fe_3C from this solution takes place at a lower temperature than the normal $\text{Ar}1$, the actual temperature depending on the concentration of the molybdenum solution. On reheating the steel after cooling from beyond the lowering temperature, the molybdenum gradually reassumes its original condition if the temperature does not exceed that producing lowering. When this change is complete the normal $\text{Ar}1$ is regained in the cooling curve.

This appears to be the only hypothesis which will satisfy the resistivity, recalescence, and carbide data, but has the apparent weakness that the iron-molybdenum compound has not been distinguished microscopically.

Taking first the resistivity data, there is no doubt that the molybdenum is not in simple solid solution in the iron, in the annealed condition. It is true that Steel 13, containing 8.01 per cent. molybdenum and 0.13 per cent. carbon, gives a resistivity which would allow 3.4 per cent. molybdenum in solution, but as the carbon increases the resistivity falls, and in the 0.9 and 1.2 per cent. carbon series the resistivity actually falls as the molybdenum content increases. This would apparently suggest that the molybdenum has reduced Benedicks' *quantité satisfaisante* to below 0.27 per cent., or that a double carbide has been formed, keeping more carbon from affecting the resistivity of the iron. The microscopic data actually shows that free carbide does exist with only 0.45 per cent. carbon in the 8 per cent. molybdenum series.

But assume that the molybdenum, not being in simple solution, is present in a state analogous to the colloidal state dispersed throughout the ferrite. The resistivity is not greatly increased, but in all probability the pearlite "saturation point" for Fe_3C will be lowered as the molybdenum concentration increases.

The resistivity of samples quenched at 800° strongly supports the view that only the Fe_3C has entered into solid solution, and note should again be taken that the 8 per cent. molybdenum steels give low values for the higher carbons accounted for largely in the microscopic examination, which shows the presence of free carbide. The resistivity values do not confirm the theory that the molybdenum exists as a double carbide, since in this case (at any rate for the steels, which show only martensite in the microstructure after quenching at 800°) the resistivity should allow for molybdenum also to be in solution. Samples quenched 1200° , described on page 36, show conclusively that at any rate a very large proportion of the molybdenum has been trapped in solid solution in the iron. The resistivity determinations therefore support the author's hypothesis in every direction.

The hardness tests prove that in annealed samples, and also on samples quenched at 800° , molybdenum does not increase the hardness (carbon being constant). Quenched from 1200° the hardness does tend to increase with increasing molybdenum and equal carbon. This supports the view that molybdenum does not form a hard double carbide, at any rate in the first two series. The carbide analyses, previously reported upon, were carried out with great care, and no definite evidence of a double carbide could be found. This is further confirmed by Carnot and Goutal,¹ who separated Fe_3Mo_2 from low carbon steels, containing respectively 2.5 and 3.4 per cent. molybdenum.

A great deal hinges on the microscopic data, and the author regrets that a special reagent has not yet been successfully employed in definitely indicating the presence of molybdenum in the structure. The fact that the pearlite is emulsified (see Steel 2, annealed, Plate XIV.) tends to support the colloidal molybdenum theory; and the fact that the "saturation" point is lowered, although favourable to the theory of the formation of a double carbide, is by no means fatal to the author's hypothesis, as already pointed out.

The specimens quenched at 800° confirm resistivity, recalescence, and hardness data in giving no indication that the

¹ *Comptes Rendus*, vol. cxxv. p. 221.

molybdenum has entered into solution in the iron or into combination with the carbon. This data is fully confirmed by the microscopic examination of quenched specimens of tungsten steels published by the author (*loc. cit.*, 1909, No. II.). The effect of the molybdenum is seen in the specimens quenched beyond the lowering temperature, in the production of the white polyhedral "austenite" structure, usually associated with high speed steels. Samples heated to 1200° and cooled down to 600° before quenching show the same characteristics as those quenched at 1200° , showing this phase to be constant down to the low recalescence under even slow cooling. Air-cooled samples from 1200° show a partial decomposition of the solid solution, giving troostite with the white austenite (or martensite). These points were also very carefully confirmed on the tungsten steels, the likeness in the air-cooled series being remarkable.

We now come to a consideration of the recalescence data in view of the different hypotheses.

The behaviour of the steels below the "lowering" temperature is strongly in favour of the view advocated, as previously described, for tungsten steels, that the change points Ac_1 and Ar_1 occurring at normal temperatures represent respectively the solution and reprecipitation of Fe_3C , and not a double carbide. The fact that the higher molybdenum steels show smaller points is surely strongly against the theory, supposing a double carbide to exist in the annealed state.

Let us consider next the theory, based on that suggested by Edwards for tungsten steels, that a carbide of molybdenum is formed beyond the lowering temperature and separates at the low point. Reheating to below the lowering temperature, and soaking, results in the reformation of iron carbide and the recovery of the normal point on cooling.

In the first place this theory, whilst supported by the author in regard to the carbon being present as carbide of iron in the annealed steel, presupposes the molybdenum to exist in solid solution. As previously stated, this is not in accordance with resistivity and other data.

Also, if the low point corresponds to the separation of a definite carbide of molybdenum, should not this point occur

at a constant temperature, at any rate over a certain range of molybdenum content? The observed data on the molybdenum steels shows the temperature of the low point to be a progressive function of the molybdenum content, and is not constant. Also, one would expect in, say, the 1 per cent. molybdenum series that a certain carbon content would be arrived at which would take up all the molybdenum to form the carbide. Beyond this the steel should show a normal Ar1 (due to excess Fe_3C) as well as the low point. No such indication has been found in the present series.

Again, it is not unreasonable to suppose that, if beyond a certain temperature the molybdenum reacts with Fe_3C to produce a carbide of molybdenum, then on cooling down and soaking at a temperature below the "lowering" point Fe_3C will be reproduced just as when the piece is first cooled and then reheated to the same temperature. Such is not the case, as shown by Curve 24 on Plate XII.

Consider also a specimen heated to 1200° and cooled fairly slowly, showing the lowered recalescence on cooling. Unless we assume that the molybdenum carbide has been decomposed during cooling—which is contrary to the original theory and is not supported by fact—then the separation of molybdenum carbide must be responsible for the low recalescence, and one would expect that the low point would be constant (or increase) in size with increasing molybdenum and constant carbon. The data actually shows that it tends to become smaller and less well defined as the molybdenum content increases.

This leads us to the question of the recovery of the normal recalescence after a low point has been produced. One of the most important facts in support of Edwards' tungsten steel theory is the necessity for soaking below the "lowering" temperature in order to bring about this recovery, this being attributed to the gradual decomposition of the tungsten carbide (assuming this to exist up to beyond 730°) and the reformation of carbide of iron. In the first place be it admitted from the special investigations on the question, both with pure tungsten-carbon and pure molybdenum-carbon steels, that slow heating and soaking is necessary for the

recovery, although, as previously noted, a series of quick reheatings will do more in this direction than either slow heating or soaking. The molybdenum steels appear to be rather more difficult to recover than tungsten steels (which on the above hypothesis would mean, of course, that the molybdenum carbide was more stable than tungsten carbide).

This theory assumes that the point of solution of molybdenum carbide is identical with that of iron carbide, since the Ac1 point occurs practically at the same temperature whether the previous cooling has shown the normal or low point. As a matter of fact, the tungsten-chrome steel upon which the theory is based showed no point on cooling from 1320° , and consequently no point on the first reheating. Subsequent reheating to 900° gradually developed Ac1 and a fairly normal Ar1.

However, examining the special curves taken on Steel 3 (Plate XII.), we find that a sample, cooled from 1200° and showing a large low point, will show a normal Ac point on reheating to 800° , and yet on cooling still fails to show a recovered normal Ar point. This strongly supports the hypothesis that the Ac point is due to the solution of carbide of iron, and that the molybdenum, not having then reverted to its original state, again prevents the reprecipitation until approximately the low-point temperature.

The author therefore suggests that the normal and lowered Ar1 points both mark the separation of carbide of iron, and the Ac point marks its solution practically independent of the condition of the molybdenum. The whole of the recalescence data is explained if, beyond this, we consider the molybdenum to undergo a certain change in its condition in solution at the "lowering" temperature, which reverts more or less slowly to its normal condition on reheating to below this temperature.

No recalescence point has been found to correspond to the "lowering" temperature, and there is no indication that a "low" point would appear in a carbonless molybdenum-iron alloy cooled from, say, 1200° . This, however, cannot be taken as a serious objection to the author's hypothesis, since the change on heating is certainly gradual, and the cooling points,

particularly in high molybdenum steels cooled from 1200° , are very broad and scattered. One would more naturally expect a recalescence at the lowering temperature if a carbide or double carbide of molybdenum were formed there.

It might be added that, in the presence of chromium, it is quite probable that a double carbide is formed, and that this is extremely stable, as demonstrated by the experiment of Carpenter¹ and Edwards (*loc. cit.*) on high-speed steels.

The author would beg to point out that the present work on molybdenum steels does not fully agree with his former theory for tungsten steels on one point, viz. in which it was suggested that the separation of Fe_3W from solution definitely occurred at the low point. This difference is not of vital importance, but the "recovery" experiments on molybdenum steels certainly tend to prove that the low point is due to Fe_3C , and it is not certain whether this is definitely connected with the reversion of the molybdenum. This point is being further investigated on steels of other molybdenum content.

Finally, let us recapitulate briefly the changes in a carbon-molybdenum steel during heating and cooling according to the author's hypothesis.

In the annealed steel the carbon exists as Fe_3C and the molybdenum as an iron-molybdenum compound, not in simple solution in the iron, but possibly as a solid colloidal solution of molybdenum or a molybdenum-iron compound in iron. If heated to anywhere below a certain temperature, called the "lowering" temperature, and cooled, the ordinary carbon changes take place, but the points are rather smaller as the concentration of this molybdenum solution increases.

The "lowering" temperature marks the commencement of a change in the state of the molybdenum, corresponding to it entering into solid solution in the iron.

The low point marks the separation of Fe_3C from the molybdenum solution, and the molybdenum may or may not revert completely to its original state, probably depending largely on the previous rate of cooling. The temperature of the low point depends on the concentration of the solid solution of molybdenum. This accounts for the gradual lowering

¹ *Journal of the Iron and Steel Institute*, 1906, No. III. p. 337.

of the point with increasing increments of temperature in high molybdenum steels (due to gradual solution of the molybdenum), and also for the limit of lowering being fixed by the molybdenum content of the steel. This theory also accounts for the fact that in two molybdenum steels of similar carbon the complete lowering of which the steel is capable is more easily obtained in the steel of lower molybdenum content. The actual temperature at which lowering commences is fairly constant in the molybdenum steels, but with tungsten steels this rises with increased tungsten content.

In certain cases the reversion of the molybdenum after producing a low point is only obtained by slow heating and soaking or repeated heating and cooling, and whilst the normal Acl point is preserved, the Ar1 point is gradually recovered. Neither the "lowering temperature" or the low point shows a recalescence point due to the molybdenum change *per se*.

In conclusion the author readily admits that further absolute proof is desirable in support of his hypothesis, but, nevertheless, in his opinion, it is the only one which will meet the facts obtained on this series of steels ranging up to 1.25 per cent. carbon and 8 per cent. molybdenum in recalescence, resistivity, carbide analysis, microscopy, and mechanical testing.

The author would here record his hearty appreciation towards Messrs. Samuel Fox & Co., Ltd., for the splendid facilities afforded him in carrying out this research, and also for permission to publish the results. His hearty thanks are due to Mr. Percy Longmuir, M.Met., for his generous help in many directions, and to Messrs. H. Robinson, S. B. Kenworthy, and J. Eastwood, members of the laboratory staff, for their patient and skilful assistance in carrying out the experimental work.

CONCLUSIONS.

The following are the conclusions based upon an examination of a series of carbon-molybdenum steels containing up to 8 per cent. molybdenum, by recalescence curves, resistivity, microscopic, and hardness data:—

Recalescence Curves.

If the initial temperature has not exceeded a certain temperature, normal points are found in cooling, independent of the rate of heating or time of soaking.

If the "lowering" temperature has been exceeded the normal Ar1 point is lowered or split, and as the initial temperature is increased assumes a definite low position.

The position of the low point is a progressive function of the molybdenum content.

The critical point is rather more easily affected as molybdenum content is increased, but a higher temperature and longer soaking is necessary to produce the full lowering effect. The temperature which first affects the point, and also that which produces maximum lowering, is slightly higher as the carbon content increases.

The low point is not disturbed by slow cooling or arrested cooling about 800° C.

The recovery of the normal point is effected by repeated reheating to below the lowering temperature. It is not easily effected by a single slow heating and soaking.

The normal Ac1 point is found on heating, and no new point has been detected corresponding to the "lowering" temperature.

Electrical Resistivity.

The influence of molybdenum on the annealed steels depends on the carbon content. In the 20 per cent. carbon series there is a slight increase with increasing molybdenum; with 0.45 per cent. carbon the resistivity is unaffected, and with 0.9 and 1.20 per cent. carbon there is a distinct decrease.

Samples quenched at 800° C. give a resistivity corresponding to carbon being in solution and molybdenum not. Quenched at 1200° C., the samples show at least a considerable portion of the molybdenum to exist in solid solution.

Microstructures.

Samples quenched at 800° C. show normal martensite, and show that the carbide is partly soluble at this temperature.

Samples quenched at 1200° C. are less easily attacked by acids, and with the higher carbon and molybdenum content show white polyhedra. Heated to 1200° C. and quenched at 600° C., the structures present essentially the same features. Samples air-cooled from 1200° show martensite and troostite in angular pattern, the proportion of the martensite increasing with the carbon and molybdenum content.

Hardness Tests.

Molybdenum does not increase the Brinell hardness in the annealed state, nor in the samples quenched from 800° C., carbon content being constant. Quenched at 1200° C., the hardness is increased, and the increase is greater with higher molybdenum.

The Constitution of Carbon-Molybdenum Steels.

Based on the above observations and on previous work on carbide analysis, &c., an hypothesis is suggested for the constitution of carbon-molybdenum steels. The data goes to prove that the carbon exists as Fe_3C in the annealed state, and is normal in its action unless the steel is heated to beyond the "lowering" temperature. The molybdenum does not exist as double carbide, and is not in solid solution in the iron, but is probably dispersed in the ferrite in a manner suggesting the existence of a solid colloidal solution of an iron-molybdenum compound in iron.

The lowering temperature marks a change in the state of the molybdenum, resulting in its behaviour as an ordinary solid solution of molybdenum (or iron-molybdenum compound) in iron, and the separation of Fe_3C is delayed until the "low" point temperature is reached. The latter depends upon the concentration of this solid solution.

The evidence of the recovery of the point by reheating to below the lowering temperature appears to show that the low point on cooling does not necessarily mean the complete return of the molybdenum to its condition in its annealed state. The exact conditions effecting the reversion of the molybdenum are not fully known, but the quickest method

of recovering the point, at any rate in Steel 3, is by repeated heating.

A theory of carbon-molybdenum steels based on the one suggested by Edwards for tungsten steels is carefully considered, but the author submits that this does not meet observed data now presented. The theory would be that in the annealed steels molybdenum exists in solid solution in iron with the carbon as Fe_3C .

Beyond the lowering temperature a carbide of molybdenum is slowly formed, and the low point is due to this carbide, and not to carbide of iron.

On the other hand, it is very probable that in molybdenum-chromium or tungsten-chromium steels double carbides are formed, and have their own distinctive properties.

APPENDIX.

THE MECHANICAL PROPERTIES OF SOME LOW MOLYBDENUM ALLOY STEELS.

In the author's previous report several interesting results in mechanical tests were obtained, which gave promise of value as medium tensile steels with good ductility, and a series of mild steels containing molybdenum and other alloys have been prepared and tested.

MANUFACTURE.

These experimental ingots have been made by the crucible process, using white clay pots, but it is assumed that in making any of the steels commercially the acid open-hearth process would be used.

The raw materials used were Swedish bar iron (W. brand), Swedish white iron (W.S. brand), 65 per cent. ferro-nickel, 65 per cent. ferro-chrome, 35 per cent. ferro-vanadium, and 85 per cent. ferro-molybdenum. The method of manufacture was the same in each case, ingots of 56 lbs. being cast into

2½-inch square moulds, and bars 1 inch round being rolled without reeling. The steels are therefore as nearly comparative as practical conditions of manufacture allow.

Steels 19 and 20 were made first, and since the carbon content was lower than was desired, the ingots were repeated under marks 23 and 24. The series therefore comprises two nickel-molybdenum, two chrome-molybdenum, one vanadium-molybdenum, and one plain molybdenum steels, whilst results of some of the earlier molybdenum steels have been included for comparison.

CHEMICAL ANALYSIS.

The following table gives the analyses of the various steels tested.

TABLE IV.—*Chemical Analysis.*

Steel No.	19.	20.	21.	22.	23.	24.
	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.	Per Cent.
Carbon	0·15	0·19	0·30	0·33	0·30	0·32
Manganese	0·22	0·23	0·33	0·28	0·24	0·28
Silicon	0·058	0·056	0·067	0·075	0·075	0·082
Sulphur	0·020	0·023	0·022	0·023	0·021	0·021
Phosphorus	0·019	0·020	0·017	0·021	0·018	0·019
Molybdenum	0·53	0·53	0·52	0·96	0·45	0·46
Nickel	0·98	0·99	...
Chrome	0·95	0·91
Vanadium	0·215

METHODS OF HEAT TREATMENT.

Twelve-inch lengths of the different steels were cut off for heat treatment, and each batch was treated at one time.

Rolled.—These tests were made on the bars as received from the mill after rolling without reeling.

Normalised.—The bars were heated for fifteen minutes at 900° C., and cooled in still air. They were cool enough to handle in about an hour and a half after withdrawing from the furnace.

Hardened and Tempered.—HT 1, HT 2, and HT 3 were quenched in whale oil from 900° C., and tempered in lead for fifteen minutes at the following temperatures:—

								Degrees.
HT 1	500
HT 2	550
HT 3	600

HT 4, HT 5, HT 6, and HT 7 were quenched in lukewarm water from 850°, and tempered in lead for fifteen minutes at the following temperatures:—

								Degrees.
HT 4	500
HT 5	550
HT 6	600
HT 7	650

In each case the bars were kept for fifteen minutes at the hardening temperature previous to quenching. After tempering, the bars were cooled in air beside the furnace wall.

MECHANICAL PROPERTIES.

From the one bar in each case, tensile alternating stress, Brinell, and scleroscope hardness tests have been made.

The Tensile Pieces were of standard section, 0.564 inch diameter, with 2 inches parallel turned from the treated bar.

The Alternating Stress Tests were made by kind permission of Professor Arnold upon the machine designed by him. A bar 6 inches long, $\frac{3}{8}$ inch diameter, is rigidly gripped to a depth of 1 inch in a die, and is struck by a plunger 3 inches above the die line. The deflection is $\frac{3}{8}$ inch on either side of the die line, and the speed 650 alternations per minute. The figures recorded represent the number of times the bar passes the vertical before fracture.

The Brinell Hardness Tests were made on sections cut from the end of the tensile piece, the surface being prepared on 00 emery. It was considered of interest to take tests on both transverse and longitudinal sections (the surface in the latter case being cut $\frac{1}{4}$ inch from the rolled surface), and the results have been plotted against the maximum stress values obtained in tensile. A Jackman machine taking 10-millimetre ball and 3000-kilogramme load was employed.

The Scleroscope Hardness Tests were made on a Shore instrument, and the author would beg to point out that these

figures are obtained on a new apparatus, and are not comparative with the tests given in his earlier report, where the diamond point was not circular, and the results were therefore low.

DISCUSSION OF TESTS.

Steel 19.—This steel is too low in carbon to expect any specially good tests, although the water-quenched and tempered samples 4, 5, and 6 all give useful figures.

Steel 20.—This steel also is low in carbon, but in the heated condition, especially No. 6, some excellent medium tonnage tests are obtained.

Steel 21.—This forms a good example of the improvement of a steel from the rolled state by heat treatment, particularly in resistance to alternating stress. The oil-quenched series shows a fairly good 50-ton steel, but the water-quenched steels, although giving a high tonnage and reduction of area, are somewhat low in elongation.

Steel 22.—Compared with the former steel, this is very stiff. Making allowances for the different size bar used in the earlier series, these tests fall well into line with Steels 1 and 2, containing respectively 0.19 and 0.44 per cent. carbon and 1 per cent. molybdenum, described in the author's first report, and appended to Table VIII. for comparison. The steel is singularly sensitive to heat treatment and to small differences in carbon content. Steel 22, giving approximately 60 tons tensile with 18 per cent. elongation, and 80 tons with 15 per cent., is worthy of further attention, although it is scarcely likely that this will compare with the nickel-chrome steels for very high tonnage properties.

Steel 23.—This steel shows useful tests in the 4, 5, and 6 treated series, giving 65 tons tensile with 18 per cent. elongation.

Steel 24.—The oil-quenched series are low in ductility, but the water-quenched Steels 5, 6, and 7 show useful values of about 80 tons tensile with 15 per cent. elongation.

Considered generally, there is no steel in this series which shows any marked superiority over the ordinary treated chromium-vanadium, or nickel, or nickel-chromium steels giving

TABLE V.—*Steel No. 19.*

Carbon, 0.15 per Cent.; Molybdenum, 0.53 per Cent.; Nickel, 0.98 per Cent.

Mark.	Tensile Test.				Hardness.				Alternating Stress.		
	Elastic Limit.	Maximum Stress.	Elongation per Cent.	Reduction of Area per Cent.	Brinell.		Scleroscope.	1.	2.	Mean.	
					Transverse.	Longitudinal.					
R	25.2	32.4	33.5	68.0	132.5	128	15.0	258	284	271	
N	22.4	32.0	33.0	62.6	135.5	128	16.7	320	320	320	
HT 1	32.52	40.48	23.5	68.0	187.0	170	24.0	190	200	195	
HT 2	32.94	41.16	27.0	70.8	192.0	197	23.5	184	196	190	
HT 3	32.4	40.48	27.0	71.6	187.0	179	25.0	196	198	197	
HT 4	37.2	44.16	24.0	67.0	217.0	202	27.5	192	176	184	
HT 5	36.84	44.20	22.5	63.6	199.0	176	23.0	178	196	187	
HT 6	37.80	45.28	23.0	66.0	228.0	202	28.5	216	190	203	
HT 7	32.6	43.6	24.5	70.0	213.0	202	24.0	174	186	180	

23

TABLE VI.—*Steel No. 20.*

Carbon, 0·19 per Cent. ; Molybdenum, 0·53 per Cent. ; Chromium, 0·95 per Cent.

Mark.	Tensile Test.				Hardness.			Alternating Stress.		
	Elastic Limit.	Maximum Stress.	Elongation per Cent.	Reduction of Area per Cent.	Brinell.		Scleroscope.	1.	2.	Mean.
					Transverse.	Longitudinal.				
R	30·6	38·4	25·0	61·6	170	140	21·0	226	210	218
N	22·2	32·0	34·5	68·0	141	128	17·0	318	296	307
HT 1	40·4	48·92	19·5	67·0	222	212	29·5	182	194	188
HT 2	41·24	49·6	21·0	66·0	238	207	32·0	168	168	168
HT 3	40·0	49·14	22·5	63·0	235	196	30·5	164	178	171
HT 4	47·8	58·08	16·0	60·4	294	255	35·0	120	140	130
HT 5	49·2	57·04	19·0	59·2	285	241	34·5	160	186	173
HT 6	49·6	57·16	20·0	63·6	289	269	34·0	142	140	141
HT 7	38·2	48·0	19·5	56·0	235	218	32·0	124	132	128

TABLE VIII.—Steel No. 22.

Carbon, 0·33 per Cent.; Molybdenum, 0·96 per Cent.

Mark.	Tensile Test.			Hardness.			Alternating Stress.		
	Elastic Limit.	Maximum Stress.	Elongation per Cent.	Reduction of Area per Cent.	Brinell.		Scleroscope.	1	2
					Transverse.	Longitudinal.			
R	40·8	47·2	16·5	49·6	232·0	212	25	176	174
N	29·36	40·4	22·0	52·4	178·5	139	22	240	252
HT 1	43·80	56·4	17·5	57·2	279·5	269	35	156	154
HT 2	49·68	59·48	18·0	54·8	281·5	269	36	148	156
HT 3	50·2	58·48	20·0	54·8	281·5	269	35	138	147
HT 4	71·28	81·64	15·5	47·2	402·0	364	50	86	85
HT 5	71·24	82·0	14·0	49·6	387·0	340	43	112	130
HT 6	73·2	82·4	15·0	47·2	364·0	351	42	96	99
HT 7	59·2	69·0	19·5	61·1	321·0	322	40	flaw	120
<i>Steel 1.—Carbon, 0·19 per Cent.; Molybdenum, 1·03 per Cent.</i>									
R	22·4	33·52	33·31	64·32	116·0
N	22·29	30·22	35·5	73·0	340	334
Oil	21·04	40·24	27·46	68·4	241·0	290	312
950/550									
<i>Steel 2.—Carbon, 0·44 per Cent.; Molybdenum, 1·05 per Cent.</i>									
R	34·4	54·4	19·5	49·23	228·0
N	37·81	52·44	20·1	49·4	387·0	180	196
Oil	75·28	94·0	14·08	49·2	150	124
900/550									

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TABLE IX.—*Steel No. 23.*

Carbon, 0.30 per Cent. ; Molybdenum, 0.45 per Cent. ; Nickel, 0.99 per Cent.

Mark.	Tensile Test.				Hardness.			Alternating Stress.		
	Elastic Limit.	Maximum Stress.	Elongation per Cent.	Reduction of Area per Cent.	Brinell.		Scleroscope.	1.	2.	Mean.
					Transverse.	Longitudinal.				
R	35.2	41.0	21.0	57.2	204.5	166	23.0	194	192	193
N	26.8	38.4	25.0	59.2	166.5	146	19.5	262	266	264
HT 1	40.6	49.28	20.0	62.6	241.0	228	33.5	160	182	171
HT 2	39.63	50.74	22.5	62.4	235.0	217	30.7	178	192	185
HT 3	38.0	48.24	23.0	60.9	235.0	217	31.7	180	208	192
HT 4	59.0	68.12	17.0	54.8	321.0	302	40.0	172	184	177
HT 5	54.9	61.6	20.0	59.2	321.0	311	40.7	150	168	159
HT 6	54.08	63.6	18.0	57.3	302.0	262	36.0	106	124	115
HT 7	40.8	55.6	20.5	60.3	228.0	217	34.0	136	134	135

TABLE X.—*Steel No. 24.*

Carbon, 0.32 per Cent. ; Molybdenum, 0.46 per Cent. ; Chromium, 0.91 per Cent.

Mark.	Tensile Test.				Hardness.			Alternating Stress.	
	Elastic Limit.	Maximum Stress.	Elongation per Cent.	Reduction of Area per Cent.	Brinell.		Scleroscope.	1.	2.
					Transverse.	Longitudinal.			
R	40.4	56.0	16.0	52.4	255	223	29	166	176
N	30.68	43.76	23.5	57.2	241	212	27	180	192
HT 1	57.76	66.13	14.5	51.6	321	312	42	134	128
HT 2	59.6	69.6	16.0	53.6	336	321	41	112	124
HT 3	55.92	65.12	18.0	51.0	311	332	39	128	125
HT 4	71.40	81.8	13.0	42.0	364	337	45	110	106
HT 5	71.0	79.5	17.5	54.8	382	364	47	131	120
HT 6	72.48	80.08	15.0	51.0	364	351	45	116	117
HT 7	52.0	69.4	18.0	62.6	321	311	41	146	146

over 50 tons tensile and over 20 per cent. elongation, although these figures are exceeded in several cases. Steel 24 gives promise of being a useful steel of a cheaper quality than the high-tonnage nickel-chromium steels now on the market, which give 110/120 tons tensile with 12/10 per cent. elongation. All the steels give notably high reduction of area figures, and

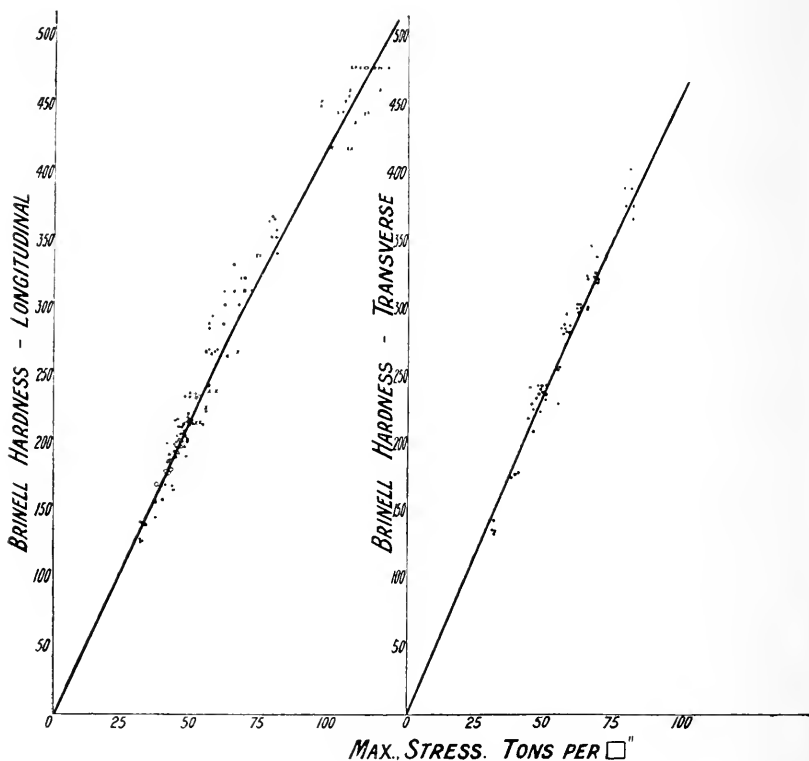


FIG. 23.—Brinell Hardness and Maximum Stress.

in this way resemble the nickel and chromium-nickel steels. An interesting feature of the alternating-stress tests is that the highest value in any set is obtained on the normalised bar, and supports the figures in the author's previous report, where it was stated, with certain reservations, that the resistance to alternating stress was in direct proportion to the ductility in static tests.

A matter of considerable interest is the correlation between the Brinell hardness and maximum stress. In Fig. 23 the author has plotted the Brinell hardness on transverse sections and on longitudinal against the stress. In the latter case the figures have been supplemented by another series of values obtained on tensile pieces of rail, spring, and treated high-tonnage steels. There is an appreciable difference between the transverse and longitudinal Brinell values, the latter being lower.

In neither case does the curve follow a straight line.

In Table XI. are set out factor values for converting approximately from Brinell hardness to maximum stress values.

TABLE XI.—*Factor Value of Brinell Hardness and Maximum Stress.*

Max. Stress.	Transverse.		Longitudinal.	
In Tensile.	Brinell.	Factor $\frac{\text{Max. Stress.}}{\text{Brinell.}}$	Brinell.	Factor $\frac{\text{Max. Stress.}}{\text{Brinell.}}$
25 tons	118	0·212	112	0·223
50 "	235	0·212	216	0·231
75 "	350	0·214	319	0·235
100 "	463	0·216	416	0·240
125 "	510	0·245

These figures, obtained by Dr. G. Dillner at the Profmingsanstalt in Stockholm on a series of carbon steels ranging from about 22 tons to 50 tons tensile, were as follows:—

(a) *With Brinell Hardness below 175—*

	Per Cent.
Coefficient for longitudinal impression . .	0·230
„ „ transverse „ . .	0·225

(b) *With Brinell Hardness above 175—*

Coefficient for longitudinal impression . .	0·219
„ „ transverse „ . .	0·206

Another interesting comparison is that of the hardness figures obtained by the Brinell and scleroscope methods. These values, obtained on one and the same specimen, are

plotted in Fig. 24, and fall about a straight-line curve, giving a factor value of

$$\frac{\text{Brinell.}}{\text{Scleroscope}} = 8.1 \text{ approximately.}$$

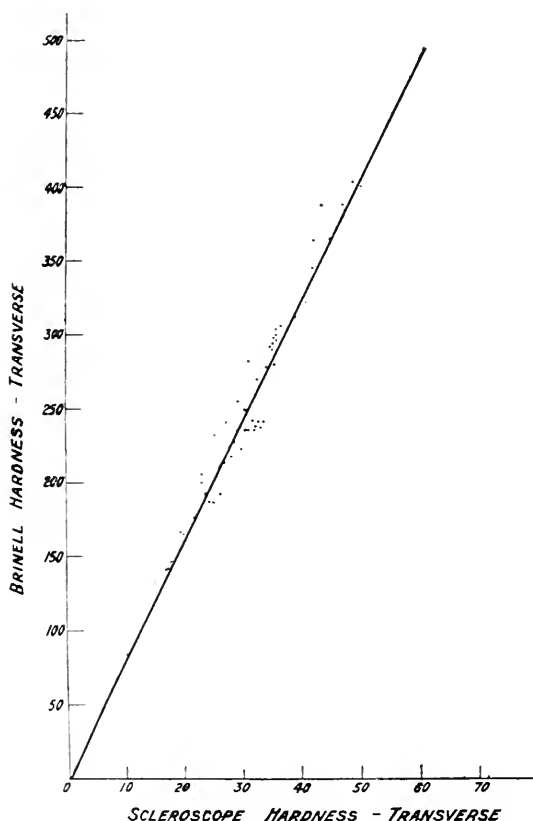
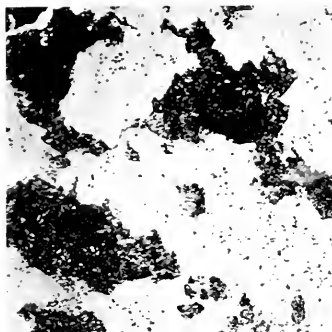
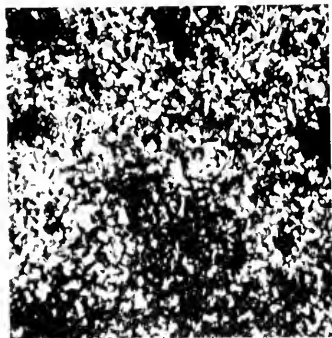


FIG. 24.—Brinell and Scleroscope Hardness.

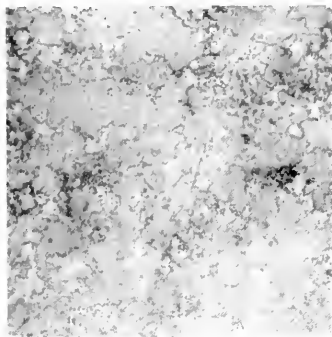
The agreement is not sufficiently near to give a definite value, but this factor is considerably above the value 6 given by the makers.



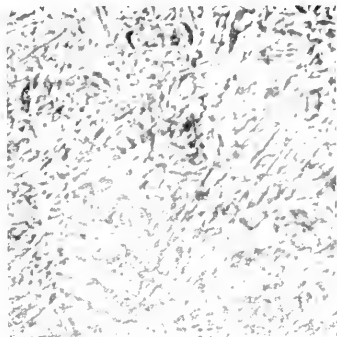
Steel 2.—Annealed.



Steel 2.—Normalised.



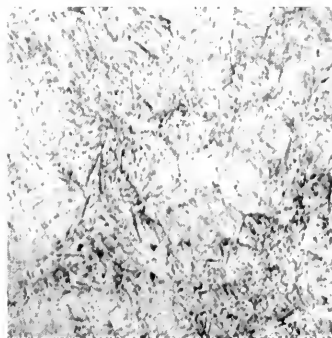
Steel 2.—Quenched at 800°.



Steel 2.—Quenched at 1200°.

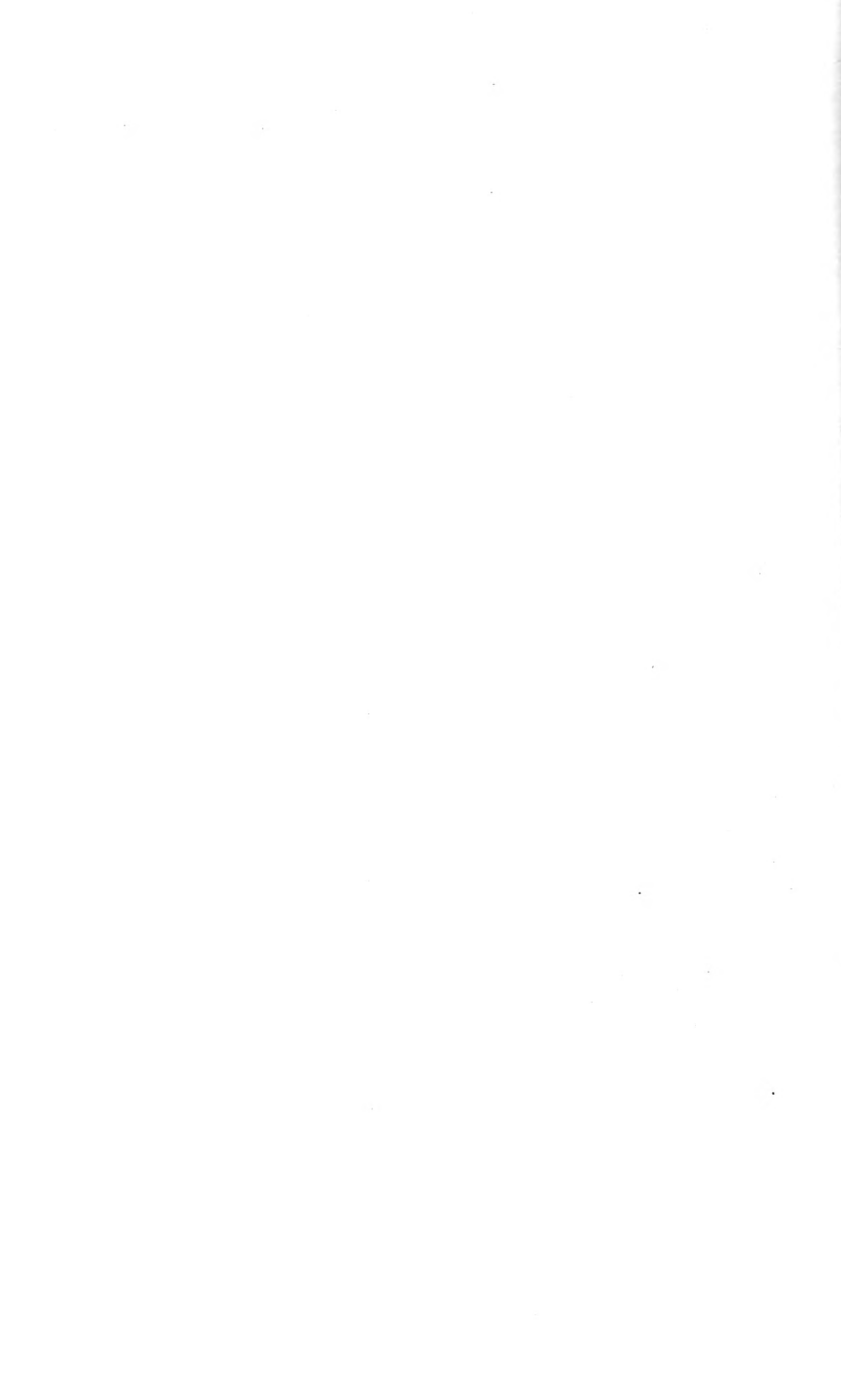


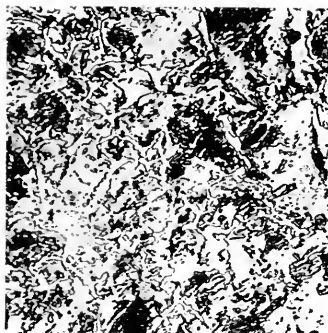
Steel 2.—Air-cooled from 1200°.



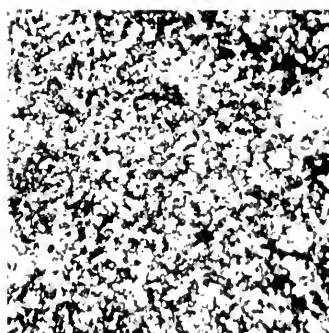
Steel 2.—Heated to 1200°, cooled to 600°, and quenched.

All magnified 150 diameters.

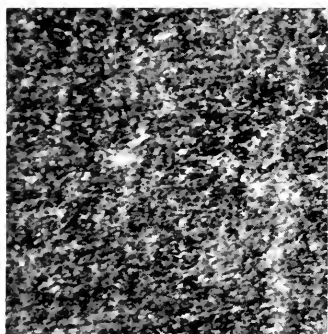




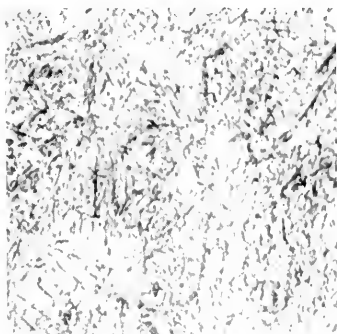
Steel 4.—Annealed.



Steel 4.—Normalised.



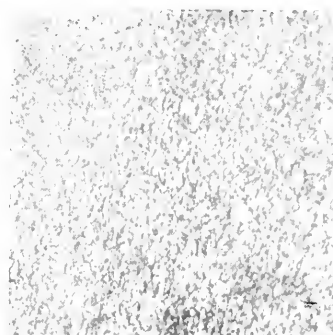
Steel 4.—Quenched at 800°.



Steel 4.—Quenched at 1200°.

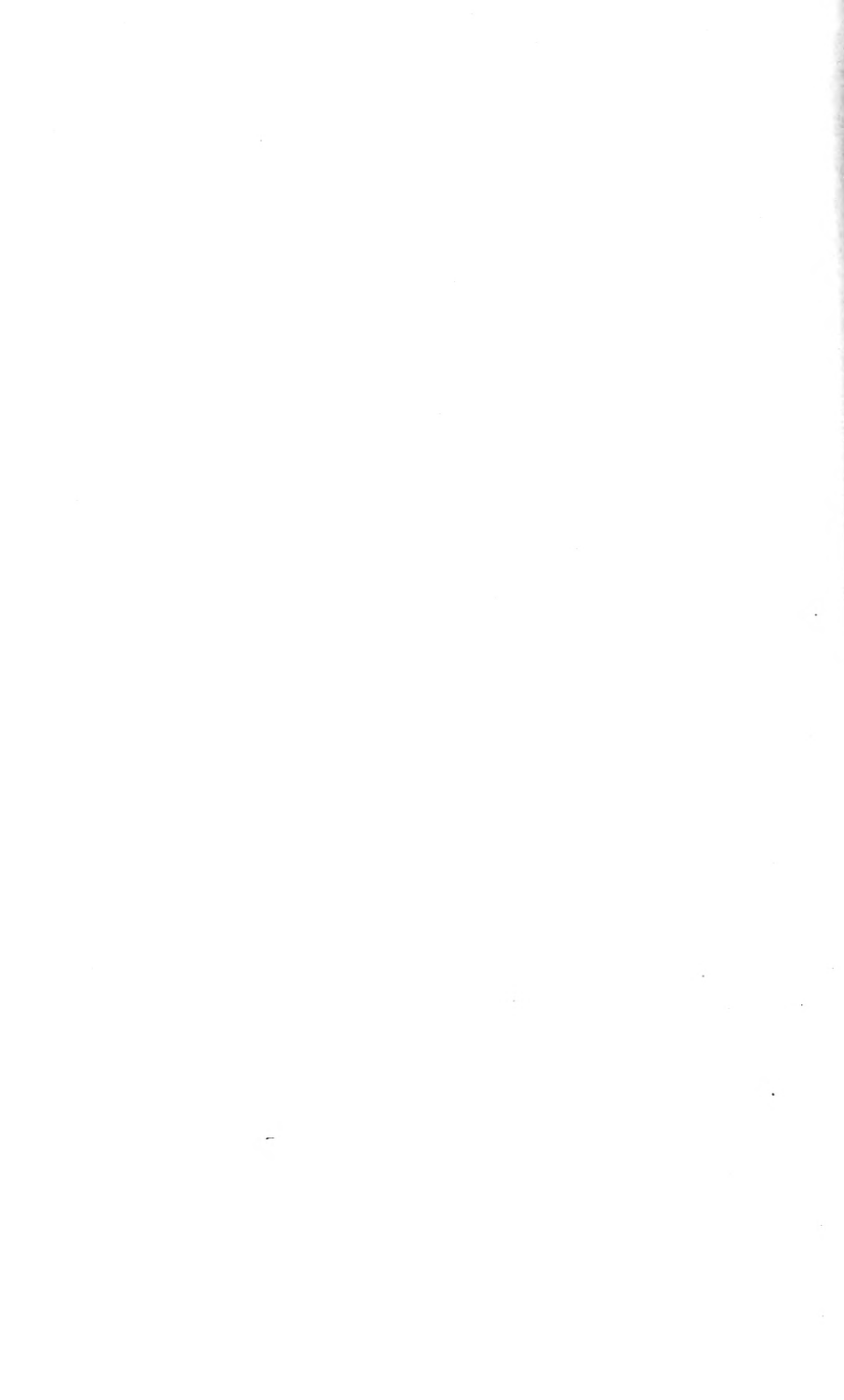


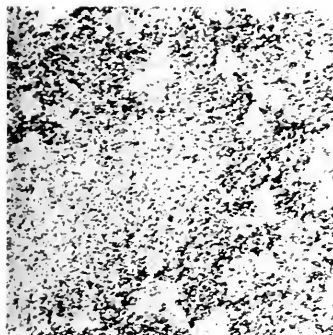
Steel 4.—Air-cooled from 1200°.



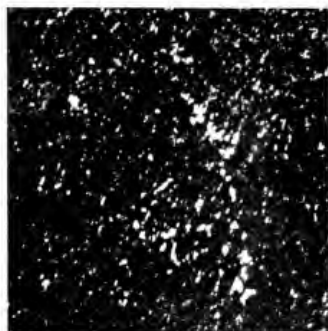
Steel 3.—Heated to 1200°, cooled to 600°, and quenched.

All magnified 150 diameters.

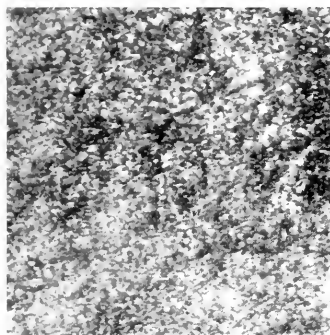




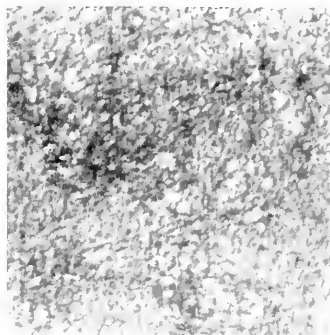
Steel 15.—Annealed.



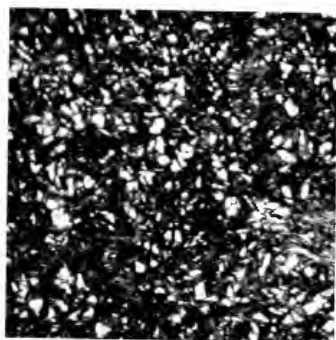
Steel 15.—Normalised.



Steel 15.—Quenched at 800°.



Steel 15.—Quenched at 1200°.

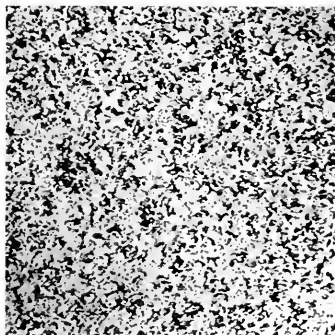


Steel 15.—Air-cooled from 1200°.

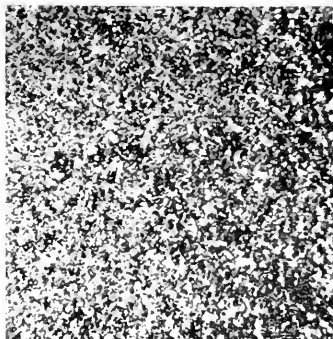


Steel 15.—Heated to 1200°, cooled to 600°, and quenched.

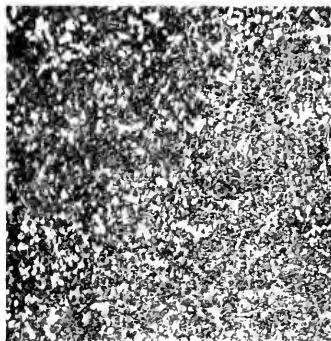
All magnified 150 diameters.



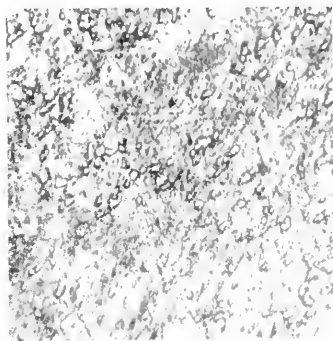
Steel 17.—Annealed.



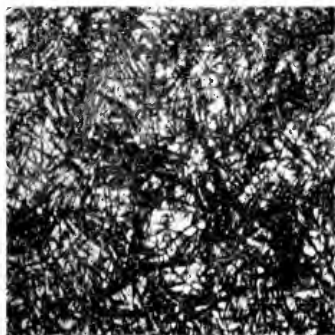
Steel 17.—Normalised.



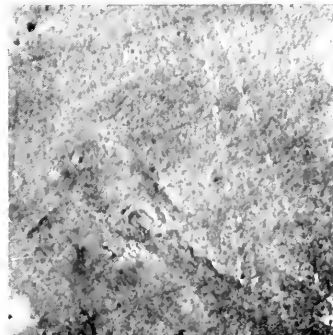
Steel 17.—Quenched at 800°.



Steel 17.—Quenched at 1200°.



Steel 17.—Air-cooled from 1200°.



Steel 16.—Heated to 1200°, cooled to 600°, and quenched.

All magnified 150 diameters.



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